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HIGH SPEED STEEL, CARBIDE SEGREGATE AND GRAIN SIZE

Edward DeMille Campbell Memorial Lecture—1936

By James P. Gill

Abstract

This lecture deals primarily with the description and results of a number of experiments made to show the characteristics and physical effects of the carbide segregate and grain size in an 18-4-1 High Speed Steel. The lecture includes a discussion on the nature and characteristics of the carbide segregate; the segregate and grain sizes as they affect the physical properties—strength and plasticity, thermal conductivity and hardness at elevated temperatures; how the carbide segregate affects the grain size; grain size as affected by repeated hardening and the grain size as affected by furnace atmospheres in hardening.

TODAY we meet for the eleventh time to commemorate the life and work of Prof. Edward DeMille Campbell. It is now about eleven years since Professor Campbell's death, sufficient time so that some of the younger members of the Society may not fully appreciate his greatness as a metallurgist, an investigator, a teacher, an administrator and a man. Dr. Zay Jeffries, in the second Campbell Memorial Lecture, briefly outlined some of his achievements and mentioned that he published about 80 papers, 65 of which were concerned with the analysis and constitution of steel. These papers displayed such careful experimental procedures and such thoughtful deductions that continued reference is made to them in our metallurgical literature.

I am deeply appreciative not only of the honor given me by the American Society for Metals but also of the interest manifested by your presence. Today I will talk on the subject of high speed steel, a material which has an economic importance far in excess of the amount produced or of its monetary value. It is a material

This is the Eleventh Edward DeMille Campbell Memorial Lecture, presented by James P. Gill, metallurgist, Vanadium-Alloys Steel Co., Latrobe, Pa. The lecture was presented at the Eighteenth Annual Convention of the Society, held in Cleveland, October 19 to 23, 1936.

of general interest, since nearly every manufacturer, regardless of the product he produces, will use at least small amounts of it. It is a complex material, so its behavior offers many opportunities for study and investigation. Today I will discuss the carbide segregate and grain size in high speed steel in an effort to explain some puzzling phenomena of its behavior.

I shall deal only with one composition of high speed steel, namely, a steel containing almost exactly 0.70 per cent carbon, 18.00 per cent tungsten, 4.00 per cent chromium, 1.00 per cent vanadium and with about 0.25 per cent of manganese and 0.30 per cent of silicon. The characteristics of a steel represented by this analysis will not only be found to varying degrees in all types of high speed steels, but apply in some degree to many types of highly alloyed steels having a medium or high carbon content. In this lecture I have purposely avoided describing such detailed experimental procedure which might become tiresome to you. This will be published with the lecture as an addenda. The lecture is divided into several parts to present it more clearly, and these will be discussed in order:

(A) The Nature and Characteristics of the Carbide Segregate.
(B) Segregate and Grain Size as they Affect the Physical Properties.
    1. Strength and Plasticity
    2. Thermal Conductivity
    3. Hardness at Elevated Temperatures
(C) Effect of Carbide Segregate on Grain Size.
(D) Grain Size as Affected by Repeated Hardening.
(E) Grain Size as Affected by Furnace Atmosphere in Hardening.

(A) The Nature and Characteristics of the Carbide Segregate

A 7-inch square ingot of 18.00 per cent tungsten high speed steel as cast and photographed at a magnification of 1000 diameters shows the characteristic appearance of photomicrograph Fig. 1, indicative of a definitely cored structure. Upon annealing to any temperature from about 1400 degrees to 1700 or 1800 degrees Fahr. (750 to 975 degrees Cent.) there results a structure as shown in photomicrograph Fig. 2, which does not show so clearly the zonal characteristics. The white parts of the skeleton-like structure and the spheroidal particles as shown in Fig. 2 are termed the carbide

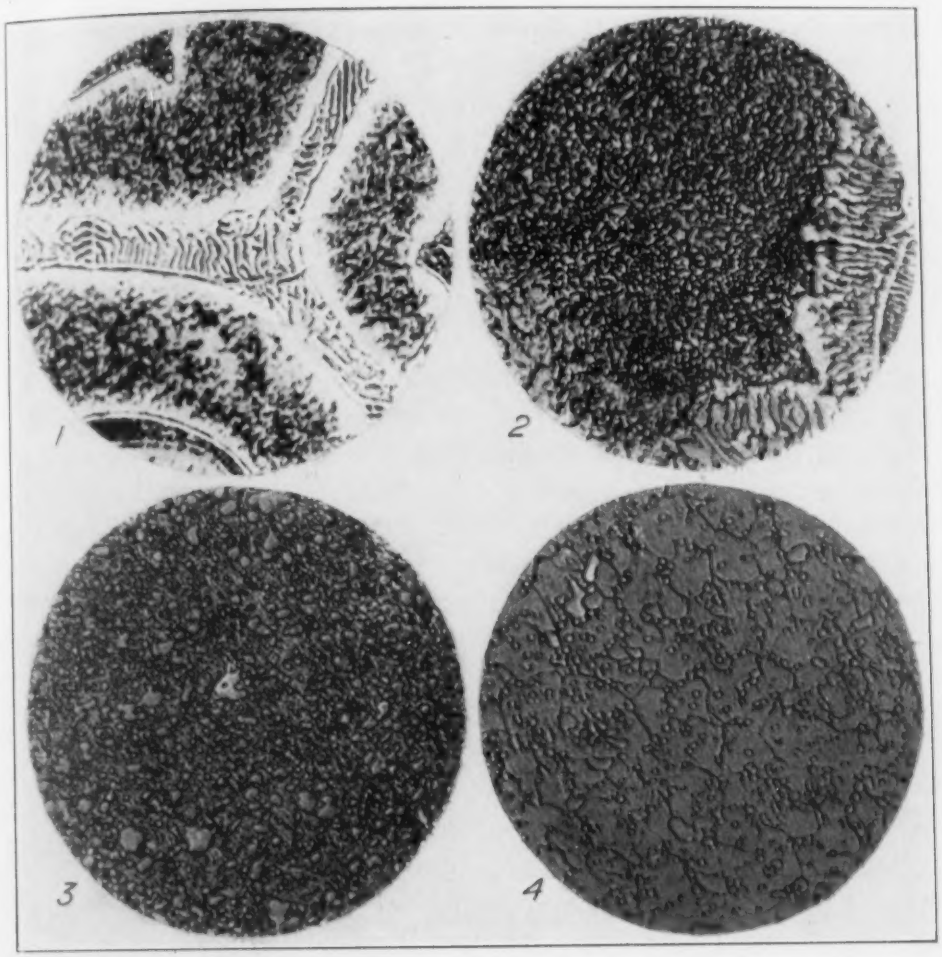


Fig. 1—0.70 Per Cent Carbon. 18-4-1 High Speed Steel Ingot As Cast. 4 Per Cent Nital. × 1000.
Fig. 2—0.70 Per Cent Carbon. 18-4-1 High Speed Steel Ingot Annealed. 2 Per Cent Nital. × 1000.
Fig. 3—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. 1½ Inch Diameter Round, Annealed. 2 Per Cent Nital. × 500.
Fig. 4—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. 1 Inch Diameter Round, Quenched from 2325 Degrees Fahr., not Tempered. 4 Per Cent Nital. × 500.

segregate. It is generally assumed that this segregate has a pronounced effect upon many of the characteristics of steels in which it occurs, but most investigators have been primarily interested in determining its composition rather than the effect it exerts upon the physical properties of the steel.

It is common knowledge that this segregate varies in size and distribution from the outside to the inside of the ingot, and that the distribution is influenced by many factors. It is also generally agreed that this segregate can only be distributed or broken up by mechanical work which causes it to elongate and break apart in the

direction of working. It is also known that there is a direct relationship in the size and distribution of this segregate when comparing the structure of the bar with the structure of the ingot from which it was made.

If a well-worked bar of an 18.00 per cent tungsten high speed steel is annealed the segregate will appear in a rounded form quite well distributed throughout the steel as shown in photomicrograph Fig. 3. There will be a number of occasions to refer to the segregate with this appearance and it will be called "spheroidal." If a bar of steel with this spheroidal structure is heated to the temperature generally recommended for hardening, there results a familiar structure as illustrated in photomicrograph Fig. 4, but by increasing the temperature slightly above that recommended for hardening, these spheroids will begin to fuse and coalesce as shown in photomicrograph Fig. 5. With further increase in time or temperature the spheroids will again, on cooling, form an eutectic-like structure considerably below the melting point of the steel, similar in appearance to that in the cast material. (Photomicrograph Fig. 6). By heating for an extended length of time at some temperature in the vicinity of 2200 degrees Fahr. (1200 degrees Cent.) or for a shorter length of time with increase in temperature, this segregate begins to have an entirely different appearance than in properly heated and annealed steel; it begins to become angular in appearance and the crystals will vary in size and degree of angularity with the time and temperature. Photomicrographs Figs. 7 and 8 show the annealed steel with the segregate in an angular appearance, exhibiting both small crystals and large crystals. Thus it is evident the segregate may have a variable distribution determined by ingot size, mold design, casting temperature, rate of cooling and by degrees of mechanical work, but the individual particles may have a variable form, either quite spheroidal or having a mixed spheroidal and angular appearance. A completely angular appearance is unusual, but when it appears it cannot be readily broken up by mechanical work. A partially angular appearance is not unusual and later in making reference to specimens having the carbide segregate wholly angular, it is for the purpose of showing how such a structure tends to affect the properties and behavior of the steel.

An 18.00 per cent tungsten high speed steel with a carbon content of 0.70 per cent can, from the unannealed condition, usually be hardened to Rockwell C 67. From the usual annealed condition

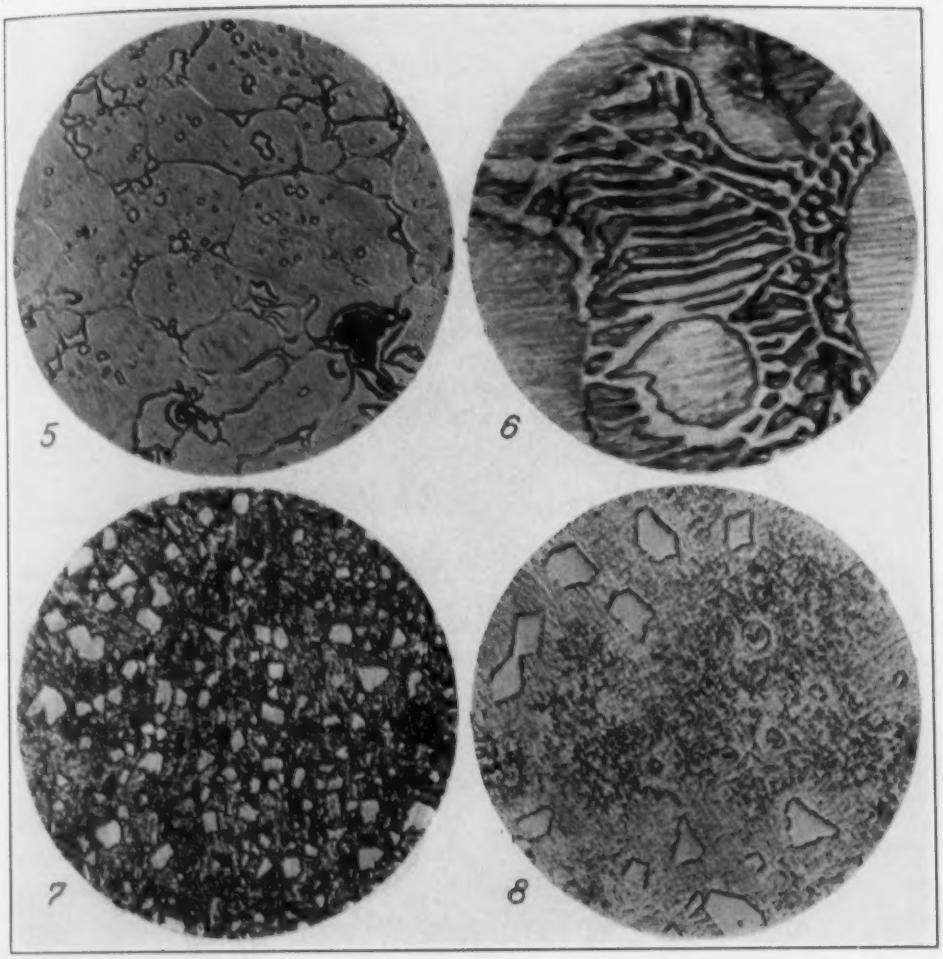


Fig. 5—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. 1 Inch Diameter Round, Quenched from 2425 Degrees Fahr., not Tempered. 4 Per Cent Nital. × 500.
Fig. 6—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. 1 Inch Diameter Round, Quenched from 2425 Degrees Fahr., Held at Temperature Longer than Specimen Fig. 5, Not Tempered. 4 Per Cent Nital. × 1000.
Fig. 7—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. 2 Inch Diameter Round, Held 8 Hours at 2300 Degrees Fahr. Annealed. 4 Per Cent Nital. × 500.
Fig. 8—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. 2 Inch Diameter Round, Held 8 Hours at 2375 Degrees Fahr. Cooled with Furnace. 4 Per Cent Nital. × 500.

it can be hardened to about Rockwell C 65-66, but if the segregate becomes angular in appearance hardenability is irregular, so that in some instances the steel can only be hardened to Rockwell C 63 and in other instances as high as Rockwell C 65, but when the higher Rockwell hardness is obtained on quenching, even tempering to 1050 degrees Fahr. (560 degrees Cent.) will lower the hardness to Rockwell C 62 or C 63. Thus it appears that the outward form of the segregate has a relationship with hardenability. This may

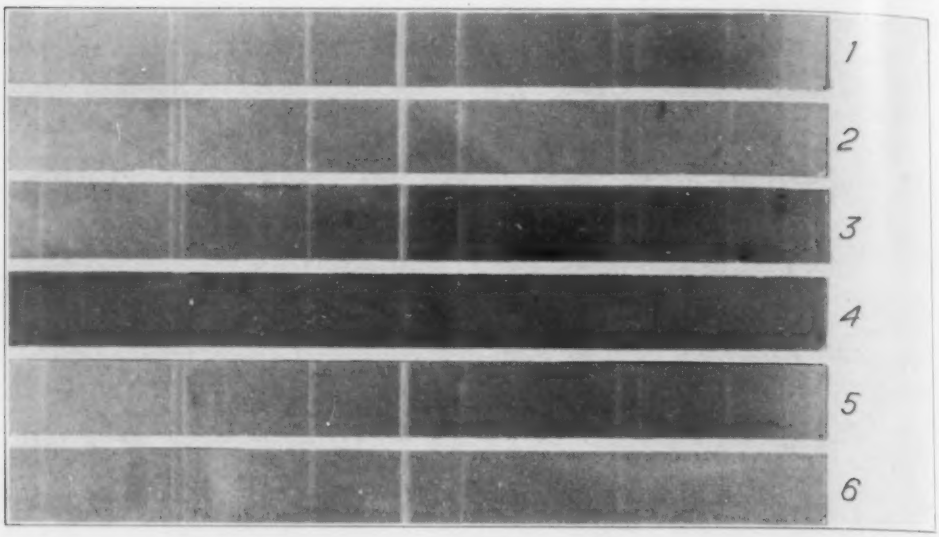


Fig. 9—Photograms of Residue of Specimens Nos. 1 Through 6 of Table I.

be because diffusion of the segregate is less rapid as it becomes more angular in appearance.

A number of investigators have displayed keen interest in the composition of this segregate. Westgren and Phragmen (1)[1] suggested the composition of Fe_3W_3C when the segregate is saturated with W, and Fe_4W_2C when saturated with Fe, and stated that it crystallized in the face-centered cubic system; they stated that a_0 equals 11.04 Ångstrom units, and that the unit cell contains 96 metal atoms and 16 carbon atoms.

By the somewhat questionable method of dissolving the iron matrix in dilute hydrochloric acid the segregate was obtained from a number of different samples treated in different ways; Table I gives the composition of the steel, its treatment, and the composition of the residue. It is interesting to note from the results that the residues from the annealed material show a higher carbon, chromium and vanadium content than the residues from the hardened samples. The differential in carbon content between the annealed and hardened samples is indicative of diffusion of the carbon from the carbides to the matrix.

The residues were also analyzed by X-ray diffraction, with the result that the photograms of all the residues given in Table I were found to be substantially the same. These photograms are given in Fig. 9. The size of the unit cell, the position of the X-ray lines

[1]The figures appearing in parentheses pertain to the references appended to this lecture.

Table I
Composition, Treatment and Composition of Residue
Same Steel Used for All Specimens

C	Si	Mn	S	P	W	Cr	Va
0.71	0.29	0.21	0.006	0.013	17.94	4.01	1.02

Specimen Number	Treatment	Composition of Residue C	W	Cr	Va
1	1-inch round bar annealed 1625 Degrees Fahr. Structure similar to micrograph number 3	2.09	63.12	4.94	2.97
2	1-inch round bar annealed and quenched 2325 Degrees Fahr.—not tempered	1.52	61.22	2.40	1.33
3	1-inch round bar annealed and quenched 2325 Degrees Fahr.—tempered 1025 Degrees Fahr.	1.47	56.74	2.67	1.50
4	1-inch round bar annealed and quenched from 2425 Degrees Fahr. Segregate fused. See micrograph number 5	1.43	61.70	2.26	1.23
5	2-inch round bar annealed. Held 8 hours at 2300 Degrees Fahr. and annealed. Segregate small angular appearance. See micrograph number 7	1.69	65.44	2.85	1.81
6	2-inch round bar annealed. Held 8 hours at 2375 Degrees Fahr. and furnace cooled. Segregate large angular appearance. See micrograph number 8	2.27	63.76	2.63	3.45
7	7-inch ingot as cast—Fig. 1	1.94	62.38	3.08	2.72
8	7-inch ingot annealed—Fig. 2	2.00	62.91	4.14	3.00

and the relative intensities of the lines are all in agreement with those observed by Westgren and his associates for the principal carbide phase in high speed steels.

In obtaining the residues given in Table I it was noted that some of the filtrates were clear while others were cloudy. By allowing the filtrates to stand for several weeks it was possible to refilter and obtain some of the residues which caused the cloudiness. The condition of the filtrate and the analysis of several of the residues recovered from the filtrates is shown in Table II. These second residues were also analyzed by X-ray diffraction and to our great disappointment gave no diffraction pattern. Since the X-ray technique was identical with that used for the original residues it is likely that the particles were of colloidal size, hence did not have sufficient resolving power to produce diffraction lines, particularly since the particles passed through filter paper. It is possible, though not likely, that the precipitate was not a primary but a reaction product.

Since it did not seem certain that only one compound existed in the residue, it was thought that possibly others existed in amounts too small to be shown by X-ray diffraction, and accordingly a number of heats were prepared with variable carbon contents and the residues obtained. The analyses of these steels and their residues are given in Table III. The residues were analyzed by X-ray diffraction and in three of the residues a new compound appears in

Table II
Condition of Filtrate and Analysis of Several Residues Recovered from the Filtrates

Specimen Number	Condition of Filtrate	Composition of Residue Recovered from Filtrate C	W
1	Very Clear		
2	Very Cloudy	0.68	64.95
3	Very Cloudy	0.41	64.5
4	Very Cloudy	....	65.0
5	Clear		
6	Clear		
7	Cloudy		
8	Clear		

addition to the one already mentioned. This compound appears in largest amount in the residue from specimen 9-X; in a lesser amount in specimen 10-X, and weakly in specimen 11-X and disappears wholly in specimen 12-X, as shown in the photograms of Fig. 10. A comparison of the diffraction pattern of this new compound with those reported by other investigators definitely shows it to be the compound Fe_2W, which is not to be expected from the suggested constitutional diagram of the iron-tungsten-carbon system.

The melting point of the carbide segregate as present in the

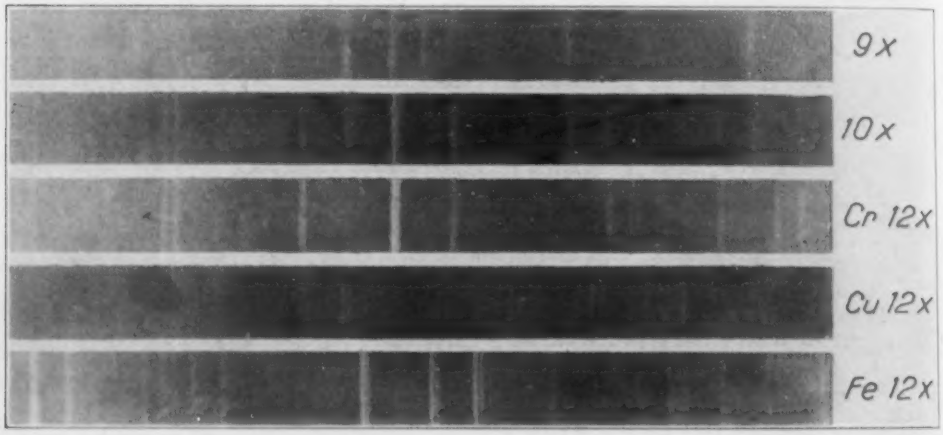


Fig. 10—Photograms of Residue of Specimens Nos. 9x, 10x and 12x. 3 Photograms of Specimen 12x Taken with Different Radiations.

Table III
Composition of Steels and Their Residues

Specimen Number	Heat Analysis C	W	Cr	Va	Condition	Residue Analysis C	W	Cr	Va
9-X	0.12	19.86	4.54	1.00	Annealed	0.42	47.02	7.91	1.51
10-X	0.22	18.16	3.79	0.98	Annealed	0.83	61.19	2.27	1.39
11-X	0.32	17.62	4.36	0.97	Annealed	1.17	62.93	3.31	1.62
12-X	0.53	17.92	3.84	1.03	Annealed	2.34	47.98	4.77	2.13

solid steel appears to be about 2400 degrees Fahr. (1310 degrees Cent.) but synthetically prepared Fe_3W_3C has a melting point much higher than this, approximately 3200 degrees Fahr. (1760 degrees Cent.). The general appearance of the carbide segregate in both the cast steel and in highly overheated steel is suggestive of an eutectic. It, therefore, appears likely that the carbide segregate is predominately composed of Fe_3W_3C, with probably small amounts of chromium and vanadium carbides, possibly a small amount of Fe_2W, and iron associated in an eutectic, which would explain the difference in apparent melting point between that observed in the solid steel and that noted for the synthetically prepared carbide. The large proportion of Fe_3W_3C present in the structure justifies the use of the word "carbide" as applied to this segregate.

The primary purpose in making X-ray diffraction patterns of these residues was not to study the basic composition of each but to make a comparison of the compositions of the residues when obtained from the steel in different structural conditions. The X-ray patterns of the residues most definitely demonstrate that regardless of the structural condition of the steel, whether it is annealed or hardened, with the segregate in a spheroidal, angular or fused condition, the primary constituent of the carbide segregate remains essentially the same.

(B) The Segregate and Grain Size as They Affect the Physical Properties

High speed steels have certain physical properties which affect profoundly the performance of the steel when used as a finished tool. The most important of these physical properties are strength, toughness, hardness at elevated temperatures, and thermal conductivity.

To show the effect of the segregate and grain size on these physical properties a number of specimens were prepared with a variable grain size and with different distributions of the segregate. The group of specimens for each type of testing were prepared with similar structures. The specimens were tested in both the as-quenched condition and after tempering for 1½ hours at 1050 degrees Fahr. (560 degrees Cent.).

Table IV shows a list of the specimens with corresponding photomicrograph Fig. numbers, structure, treatment and hardness.

Table IV

Composition of Steel—Same for All Specimens

C	Si	Mn	S	P	W	Cr	Va
0.71	0.26	0.24	0.005	0.016	18.02	3.98	1.01

Specimen Numbers	Fig. Numbers	Structure	Treatment	Rockwell Hardness—C Scale As-Quenched	Rockwell Hardness—C Scale Tempered 1050°F.
11	Fig. 11	Small grain. Small spheroidal segregate.	Preheat 1550 Degrees Fahr. High heat 2325 Degrees Fahr.	66	64
12	Fig. 12	Medium size grain. Small spheroidal segregate.	Preheat 1550 Degrees Fahr. High heat 2325 Degrees Fahr. Held in high heat 3 minutes longer than specimen 11.	66	65
13	Fig. 13	Large grain. Small spheroidal segregate.	Preheat 1550 Degrees Fahr. High heat 2325 Degrees Fahr. Re-treated from high heat 2350 Degrees Fahr.	65	65
14	Fig. 14	Large and medium grain mixed. Small spheroidal segregate.	Preheat 1550 Degrees Fahr. High heat 2350 Degrees Fahr. Treatment repeated 2 times.	66	65
15	Fig. 15	Small to medium size grain. Small spheroidal segregate.	Annealed 1600 Degrees Fahr. and given original treatment.	67	65
16	Fig. 16	Medium size grain. Small spheroidal segregate.	Specimen not annealed after rolling—Preheat 1550 Degrees Fahr. High heat 2350 Degrees Fahr.	66	65
17	Fig. 17	Slight fusion of segregate at grain boundaries. Medium to large grain. Segregate small with angular appearance.	Preheat 1550 Degrees Fahr. High heat 2350 Degrees Fahr. Stress relieval treatment. Hardened second time as original.	65	65
18	Fig. 18	Segregate with angular appearance—small crystals. Medium size grain.	Preheat 1550 Degrees Fahr. High heat 2425 Degrees Fahr. Rapid heating.	65	63
19	Fig. 26 Fig. 27	Segregate spheroidal. Highly segregated. Specimen from center of 8 inch square billet from 14 inch ingot.	Heated to 2300 Degrees Fahr. 8 hours. Stress relieval treatment—hardened preheat 1550 Degrees Fahr. High heat 2350 Degrees Fahr.	66	64
20	Fig. 20	Small to medium size grain. Segregate spheroidal. Highly segregated but less so than specimen 19. Specimen 1 inch from outside 8 inch square billet from 14 inch ingot. Small grain.	Preheat 1550 Degrees Fahr. High heat 2350 Degrees Fahr.	66	65

All specimens from 10 inch square ingots forged to small billets and hot rolled to a size slightly larger than the specimen, unless otherwise noted. All specimens given an anneal at 1635 degrees Fahr. before treatment unless otherwise noted.

Specimens when tempered have the letter "T" after the number. Tempering when performed was at 1050° F. for 1½ hours at temp.

Small grain refers to about #8 plus, medium about #7 and large about #1 A.S.T.M. Standard but both small and medium would be 9½ to 10 Shepherd fracture Standard.

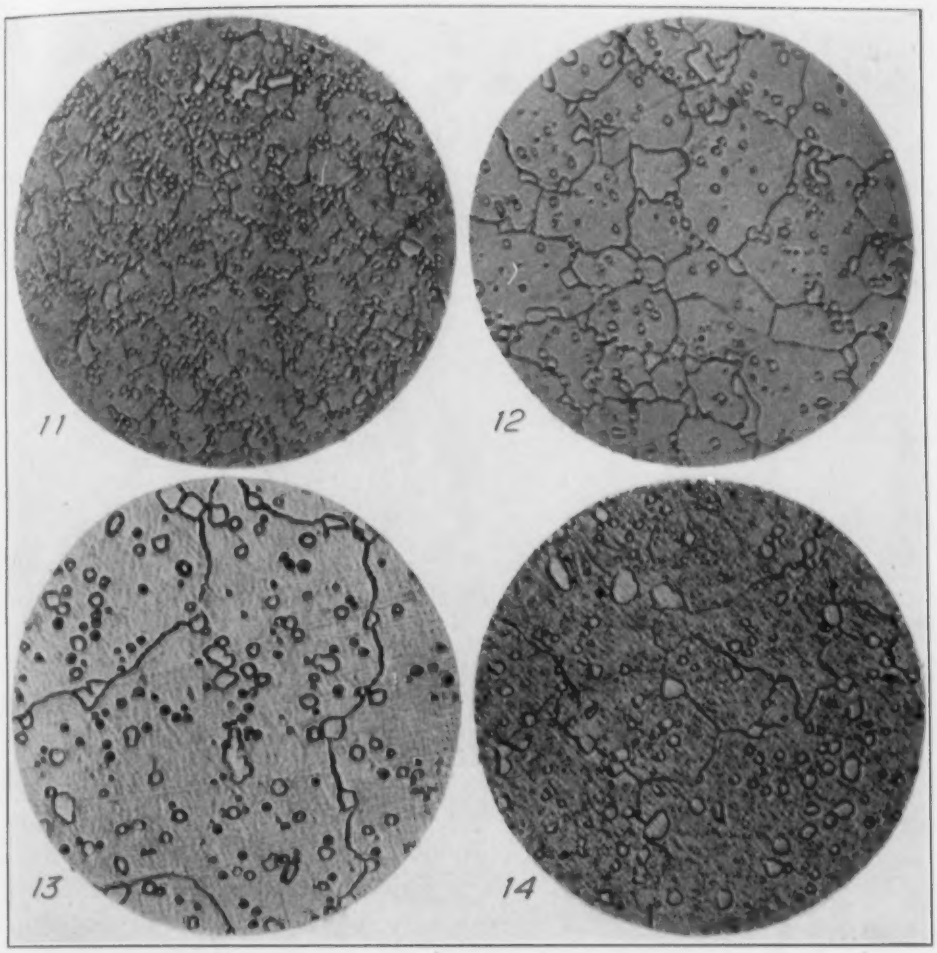


Fig. 11—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Quenched from 2325 Degrees Fahr. Not Tempered.

Fig. 12—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Quenched from 2325 Degrees Fahr. in High Heat 3 Minutes Longer than Specimen Fig. 9. Not Tempered.

Fig. 13—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Original Quench 2325 Degrees Fahr. Requench 2350 Degrees Fahr. Not Tempered.

Fig. 14—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Quenched from 2350 Degrees Fahr. Requench from Same Temperature. Annealed at 1600 Degrees Fahr. and Quenched from 2350 Degrees Fahr. Not Tempered. All Specimens Etched in 4 Per Cent Nital. × 500.

The specimens are numbered according to structure, and reference is made to these specimen numbers in the determination of the different physical properties. In most instances the grain size was obtained in a single treatment by varying the time and temperature, thus also affecting the amount of the segregate that would be dissolved in the matrix. In referring to the effect of grain size on the physical properties, the collateral effect of a different solution of the segregate in the matrix must be taken into consideration.

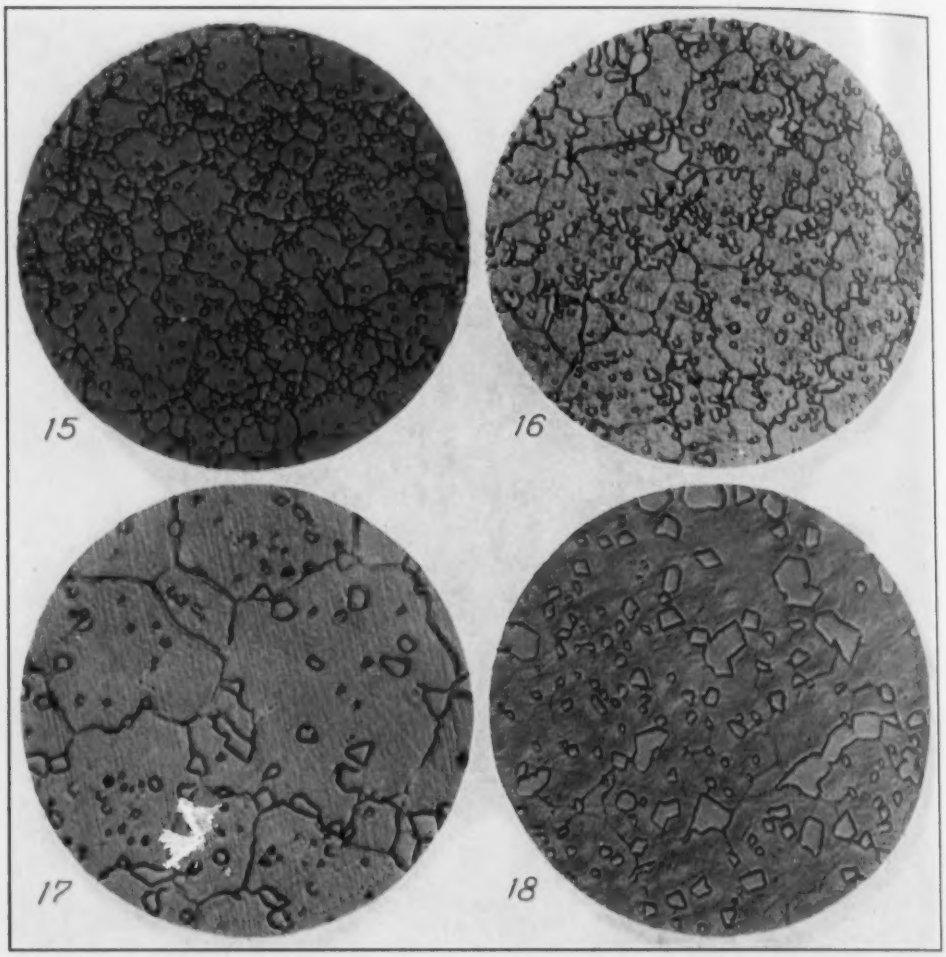


Fig. 15—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Not Annealed After Rolling. Quenched from 2350 Degrees Fahr. Not Tempered.

Fig. 16—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Quenched from 2350 Degrees Fahr., Stress Relieving Treatment, Quenched from 2350 Degrees Fahr. Not Tempered.

Fig. 17—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Slight Fusion of Segregated Grain Boundaries. Quenched from 2425 Degrees Fahr. Rapidly Heated. Not Tempered.

Fig. 18—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Segregate with Angular Appearance. Held 8 Hours at 2300 Degrees Fahr. Stress Relieving Treatment, Quenched from 2350 Degrees Fahr. Medium Grain Size. Not Tempered. All Specimens Etched in 4 Per Cent Nital. × 500.

(B-1) Strength and Plasticity

The strength and plasticity of high speed steels can be varied over a wide range by only slight changes in grain size and in segregate distribution. These physical properties are quite difficult to determine accurately, and a study of the several methods that have been proposed for determining either or both indicates that the method of torsion testing as proposed by Emmons (2) and described

by him in a paper before the American Society for Testing Materials in 1931, could be used with the greatest degree of accuracy as a comparative method of determining these properties. Emmons proposed test specimens as shown in Fig. 20, which specimens were broken in torsion. Fig. 21 is a typical illustration of the method of using the torsion test data. The torque in inch-pounds is plotted

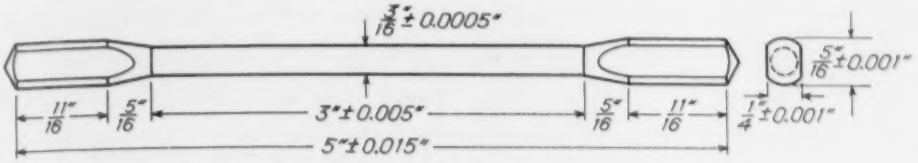


Fig. 20—Sketch of Torsion Test Specimen.

as the ordinate and the angular deformation or twist in degrees is plotted as the abscissa. The distance *AB* Emmons termed the elastic deformation and the distance *BC* the plastic deformation. Strength within the elastic limit would be indicated by the distance *DE*, and ultimate strength by the distance *CF*.

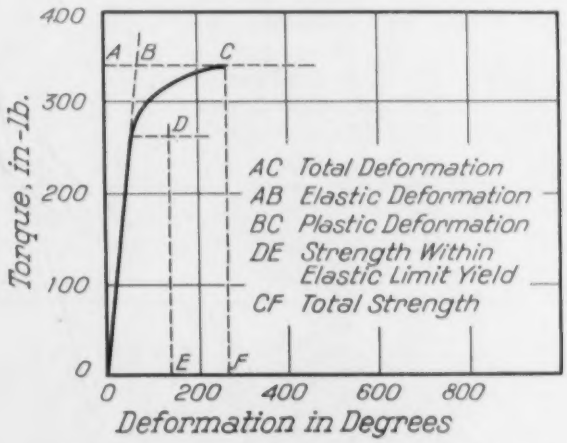


Fig. 21—Typical Illustration of the Method of Using Torsion Test Data.

Specimens were prepared and tested in triplicate and the average of the results plotted. Since straightening cannot be performed on the specimens after hardening without seriously affecting the results, the specimens which did not remain straight within 0.005 inch after hardening were discarded and additional specimens prepared.

All specimens were made from hot-rolled bars, 3/8 inch diameter, produced from the same ingot and had a composition of almost exactly 0.70 per cent carbon, 18.00 per cent tungsten, 4.00 per cent

chromium and 1.00 per cent vanadium. The specimens were then treated to obtain the desired structure.

Fig. 22 shows the stress-strain curves of specimens treated in several ways and then tempered to 1050 degrees Fahr. (560 degrees Cent.) for 1½ hours. Specimen 11-T had a small grain; specimen 12-T a medium grain and specimen 13-T a large grain. A comparison of the strength and plasticity of these three specimens shows a direct relationship to grain size, the plasticity decreasing at a very rapid rate with increase in size of the grain.

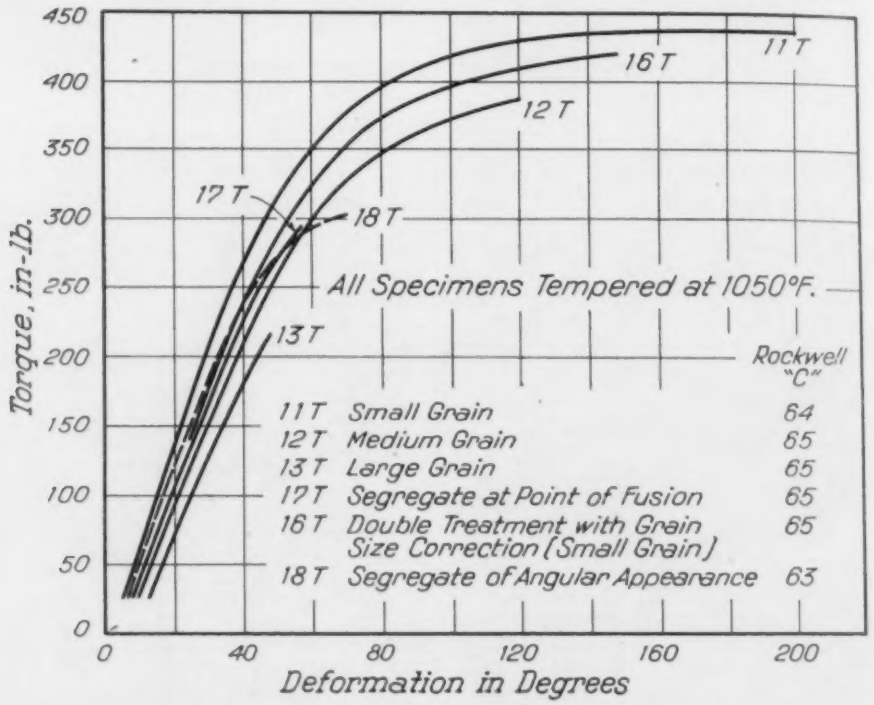


Fig. 22—Stress-Strain Curves of Specimens Treated in Several Ways and Then Tempered at 1050 Degrees Fahr. for 1½ Hours.

Specimen 16-T was given the same original treatment as specimen 11-T, then a stress relief treatment (which will be described later) and a second hardening the same as the original. It had a grain size substantially the same as specimen 11-T but appears definitely to have lost some of its plasticity by the double treatment. Specimen 17-T contained a segregate which had been near the fusion point and, as would be expected, shows substantially no plastic deformation.

Specimen 18-T was treated to produce the segregate with a crystalline appearance and it is very brittle—more so than can be

explained on the basis of the grain size of this sample. It would appear that the sharp corners and edges of the segregate in this form had a wedge-like effect, thereby making the material much less plastic. It is not improbable that another contributing factor toward embrittlement might be that the bond between the segregate and matrix is weakened by a lesser tendency of the angular segregate to diffuse than when the segregate is in a rounded form.

In Fig. 23 are shown similar specimens to those in Fig. 22 but

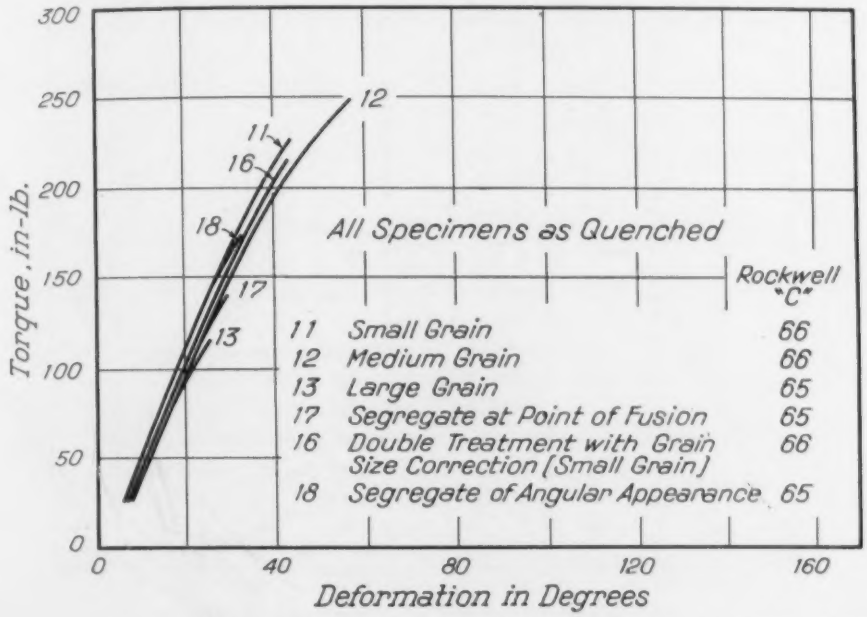


Fig. 23—Similar Specimens to Those of Fig. 22 But in the As-Quenched and Not Tempered Condition.

in the as-quenched and not tempered condition. These specimens show almost no plastic deformation, undoubtedly due to the stresses that remain after quenching, but even in this condition the specimens show a relative strength very similar to that in Fig. 22 after tempering. In Fig. 24 is shown a comparison of specimens heat treated (quenched and drawn) from the as-rolled and annealed conditions. The difference in physical properties between specimens 11-T and 15-T is not large but favors slightly the specimen treated from the as-rolled condition.

In Fig. 25 is shown the effect of the condition of the segregate on the physical properties; the segregate being in the spheroidal form but existing in large masses. Specimens 19 and 19-T are from the center of an 8-inch square billet, microstructure of which is

shown in the as-quenched condition in photomicrographs Figs. 26 and 27. Specimens 20 and 20-T were also from an 8-inch square billet, but taken at a depth of 1 inch from the outside surface, the microstructure of which is shown in photomicrograph Fig. 28. A comparison of the plasticity of these specimens with that of a speci-

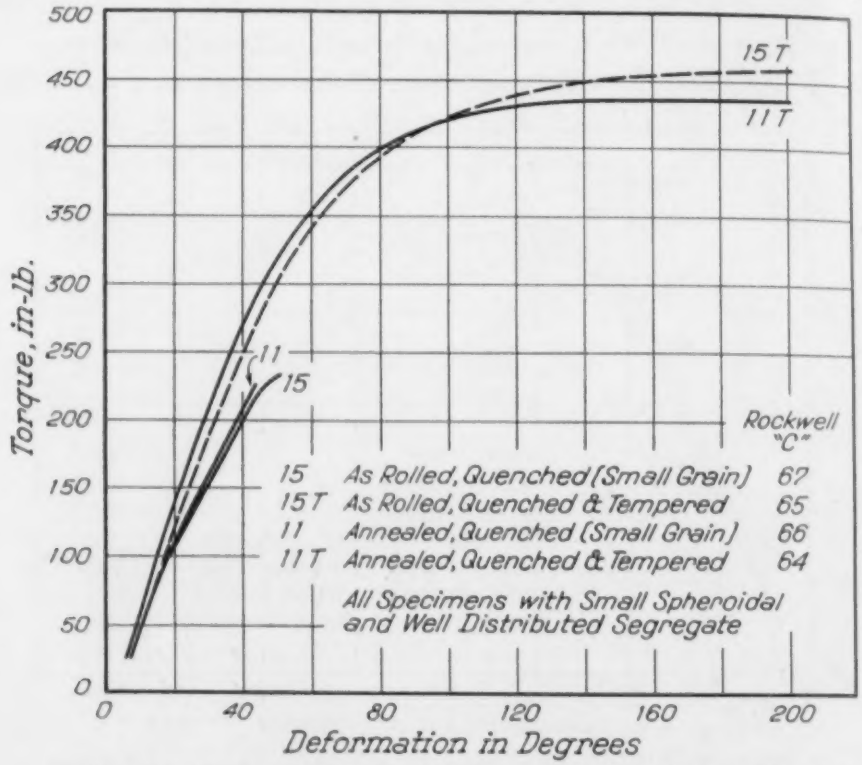

Torque, in-lb.
Deformation in Degrees
15 T
11 T
11
15
Rockwell "C"
15 As Rolled, Quenched (Small Grain) 67
15 T As Rolled, Quenched & Tempered 65
11 Annealed, Quenched (Small Grain) 66
11 T Annealed, Quenched & Tempered 64
All Specimens with Small Spheroidal and Well Distributed Segregate

Fig. 24—Shows a Comparison of Specimens Heat Treated (Quenched and Drawn) from the As-Rolled and Annealed Conditions.

men having the segregate in a well distributed form (11-T) indicates the amazing effect that the distribution of the segregate has upon the physical properties.

It is evident from these physical tests that grain size, distribution and form of segregate affect profoundly the strength and plasticity of high speed steels. The strength and plasticity was considerably lower in the specimens having the segregate in angular form. The tendency for the segregate to assume this form is somewhat proportional to the temperature and time at temperature to which the material is subjected; thus, long heating for any purpose may affect the physical properties in the final treated product. Even annealing appears to tend to decrease these properties slightly. Thus, homogenizing treatments given to high speed steel, essentially

for the purpose of breaking up the cored structure, will tend to place the segregate in an angular form, thereby inducing brittleness or lack of plasticity.

(B-2) Thermal Conductivity

The thermal conductivity of high speed steel is of importance, since the failure of a tool is often the result of the softening of the cutting edge from the temperature to which it may be heated; the

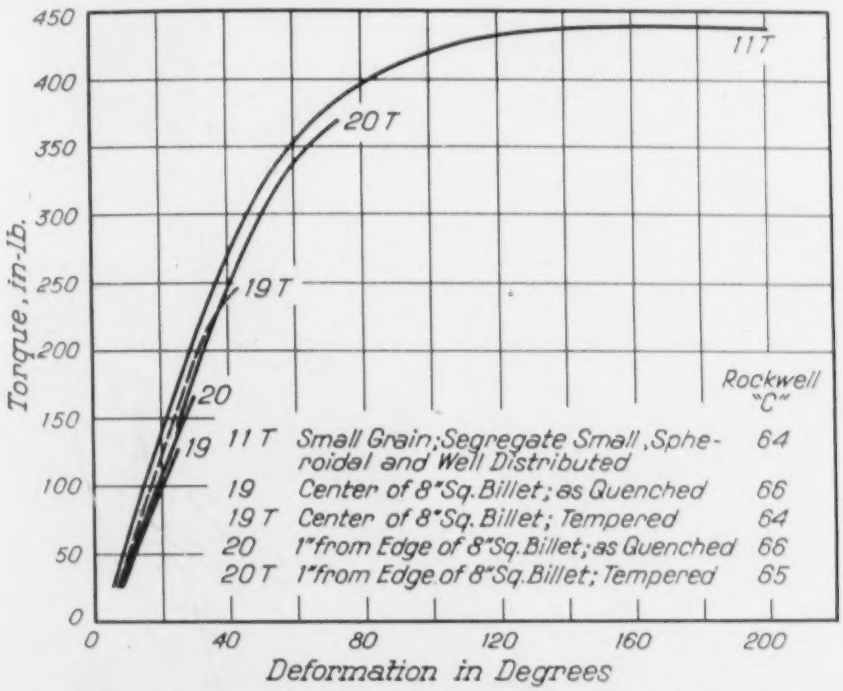


Fig. 25—Curves Showing the Effect of the Condition of the Segregate on the Physical Properties; the Segregate Being in a Spheroidal Form but Existing in Large Masses.

rapidity with which the steel will conduct heat from the cutting edge will have a considerable effect upon the rise in temperature at the cutting edge, which in turn affects tool life.

The only important study ever made on the thermal conductivities of tool steels is that made by Hattori (4) and published in 1934 in the *Journal* of the Iron and Steel Institute (British). Hattori found that high speed steel, quenched and tempered, showed about 35 per cent greater thermal conductivity than when the material was not tempered. This large difference in thermal conductivity between tempered and untempered high speed steel is an important explanation of the much longer life of high speed steels

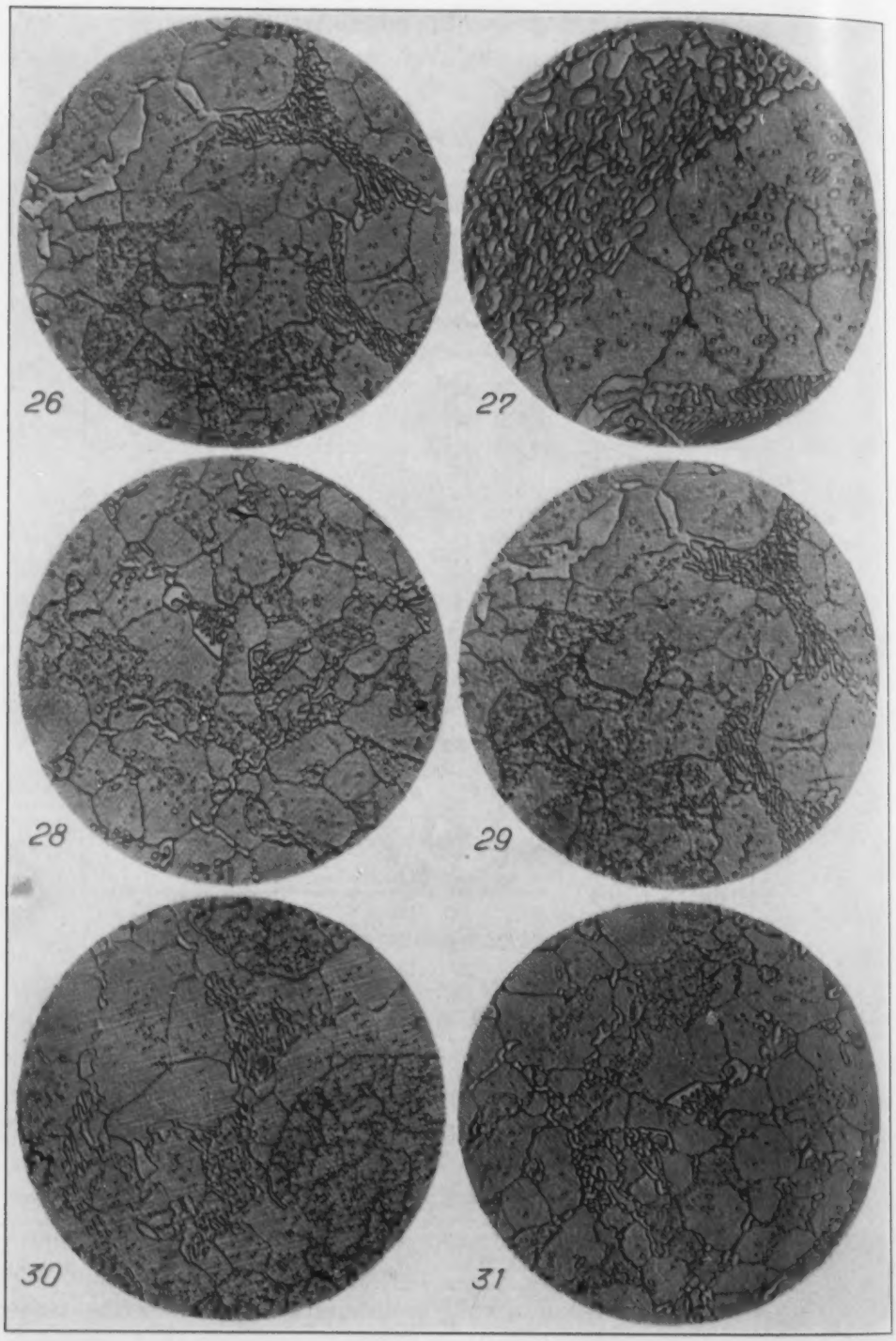


Fig. 26-31—Photomicrographs of 0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Fig. 26—From Center of 8 Inch Square Billet. Quenched from 2350 Degrees Fahr. Not Tempered. Fig. 27—Longitudinal Section of Fig. 26. Fig. 28—One Inch from Outside of 8 Inch Square Billet. 2350 Degrees Fahr. Quenched, Not Tempered. Fig. 29—From Center of 8 Inch Square Billet. Quenched from 2350 Degrees Fahr. Not Tempered. Fig. 30—4 Inches from Center of 8 Inch Square Billet. Treatment Same as Fig. 29. Fig. 31—One Inch from Outside of 8 Inch Square Billet. Treatment Same as Fig. 29. All Specimens Etched in 4 Per Cent Nital. × 500.

Table V

Composition of Steel—Same for All Specimens

C	Si	Mn	S	P	W	Cr	Va
0.71	0.26	0.24	0.005	0.016	18.02	3.98	1.01

All values are given in calories per centimeter cube per degree Centigrade at 20 to 30 Degrees Cent.
Tests were made with a temperature differential of 55 Degrees Cent.
Size of Samples—0.375 × 0.200 × 0.750 in.
All specimens quenched and tempered.
See Table IV for more detailed treatment.

Specimen Numbers	Micro Numbers	Structure	Rockwell Hardness C Scale	Thermal Conductivity
11	11	Small grain. Small spheroidal segregate.	64	0.066
12	12	Medium size grain. Small spheroidal segregate.	65	0.068
13	13	Large grain. Small spheroidal segregate.	65	0.063
14	14	Large and medium grain mixed. Small spheroidal segregate.	65	0.062
15	15	Small to medium size grain. Small spheroidal segregate.	65	0.064
17	17	Slight fusion of segregate at grain boundaries. Medium to large grain. Segregate small with angular appearance.	65	0.064
18	18	Segregate with angular appearance. Large crystals. Large size grain.	63	0.062
19	19	Segregate spheroidal. Highly segregated. Specimen from center of 8 inch square billet from 14 inch ingot. Small to medium size grain.	64	0.061
20	20	Segregate spheroidal. Highly segregated but less so than specimen 19. Specimen 1 inch from outside 8 inch square billet from 14 inch ingot. Small grain.	65	0.065
21	1	As cast. 2350 Degrees Fahr. quench. 1050 Degrees Fahr. T.	64	0.067

tempered to about 1050 degrees Fahr. (560 degrees Cent.) in contrast to that when the steel has not been tempered. The thermal conductivity of alloys, like the electrical conductivity, decreases sharply with solid solution formation; thus it would not seem likely that this property would be materially affected by change in grain size or by distribution and form of the segregate. The importance of this property made the possibility of such effect seem worthy of investigation.

Specimens were prepared in a number of different manners as already described and given in Table IV. The thermal conductivity was then determined on each of these specimens, with the results as shown in Table V. The values given in this table for thermal conductivity are accurate only for purposes of comparison, and not in an absolute sense. Specimens tested in the as-quenched condition and not tempered did not show results comparable to the work of

Hattori, possibly owing to the fact that the test procedure of Hattori was more accurate than that which was followed here. The specimens as quenched and not tempered did, however, show definitely lower thermal conductivity than in the tempered condition. A comparison of the thermal conductivity of the samples having different grain size and different distribution and form of the segregate would indicate that these variations in structure do not materially affect the thermal conductivity and that, as expected, the thermal conductivity is chiefly affected by the amount of solid solution present in the structure of the steel.

(B-3) Hardness at Elevated Temperatures as Affected by Grain Size and Segregate

A most important physical property of high speed steel is its ability to retain its hardness at elevated temperature, the so-called "red hardness" or "hot hardness." A number of efforts have been made to determine this property of high speed steel, which as a whole were rather unsatisfactory until the investigations of Harder and Grove (5). These workers made an excellent study of testing procedure and conducted a large number of tests of high speed steels on different compositions, following primarily the test procedure of Cowdrey, (6) known as the "mutual indentation method of hardness testing." The results were presented in a paper (5) at the New York meeting of the American Institute of Mining and Metallurgical Engineers in February, 1933. Their conclusions were that the hardness of high speed steels at elevated temperatures show a definite relation to the cutting properties of the steels when used in production.

Since the work of Harder and Grove seems to attach much significance to the property of "hot hardness," a number of specimens were prepared of exactly the same composition of an 18-4-1 high speed steel in accordance with Table IV. The specimens were prepared to the exact size suggested by Harder and Grove and were tested at the Battelle Memorial Institute, using the identical equipment and procedure as described in the paper mentioned.

Most of the specimens were tested in duplicate and were tested in both the quenched and tempered condition. The specimens were all tested at the same temperature, namely, 1200 degrees Fahr. (650 degrees Cent.) with the results shown in Table VI.

Table VI

	C	Si	Mn	S	P	W	Cr	Va
Composition of Steel—	0.71	0.26	0.24	0.005	0.016	18.02	3.98	1.01

Size of Specimens—1 cm. diameter × 1 cm. long—ground finish.
Tested at 1200 Degrees Fahr.

Specimen Number	Micro Number	Final Treatment	Structural Condition	Average Brinell Hardness by Mutual Indentation
11	11	Quenched—not tempered.	Small grain.	338
11-T		Tempered.	Small spheroidal segregate.	345
12	12	Quenched—not tempered.	Medium size grain.	385
12-T		Tempered.	Small spheroidal segregate.	395
13	13	Quenched—not tempered.	Large grain.	406
13-T		Tempered.	Small spheroidal segregate.	415
14	14	Quenched—not tempered.	Large and medium size mixed grain.	384
14-T		Tempered.	Small spheroidal segregate.	375
15	15	Quenched—not tempered.	Small to medium grain—specimen quenched as rolled.	385
15-T		Tempered.	Small spheroidal segregate.	355
16	16	Quenched—not tempered.	Medium size grain — double treatment with stress relieval treatment between.	381
16-T		Tempered.	Small spheroidal segregate.	406
17	17	Quenched—not tempered.	Slight fusion of segregate at grain boundaries.	415
17-T		Tempered.	Large grain.	406
18	18	Quenched—not tempered.	Segregate with angular appearance.	335
18-T		Tempered.		388

It is most interesting to note that the "hot hardness" increased directly with increase in grain size; the smallest grain showing a Brinell hardness average for both quenched and drawn specimens of 342; the medium grain an average Brinell hardness of 390 and the very large grain an average Brinell hardness of 410. The specimen which was quenched from the point where the segregate is just beginning to fuse in the grain boundaries, which also had a very large grain, is substantially of the same hardness as the specimen treated to produce a large grain without fusion of the segregate. The specimen which was quenched from the as-rolled condition and which shows a grain size intermediate between very small and medium, shows an average hardness of Brinell 369, exactly where it should be placed in accordance with its grain size. The specimen which was quenched four times from 2350 degrees Fahr. (1285 degrees Cent.) had a mixed grain size although, in general, the grains were very large; its average hardness was Brinell 380. The specimen hardened once, given a grain size correction and re-hardened had a medium sized grain and showed an average hardness of Brinell 394. This is consistent in hardness with the specimen given

a single quench and having a comparable grain size. The specimen having an angular segregate was unique, for it showed an average hardness of Brinell 335 in the as-quenched condition—the lowest of any of the specimens—but in the tempered condition a hardness of Brinell 388.

The results of these tests seem to be consistent with cutting tests, for so long as chipping of the cutting edge does not take place as a result of the brittleness induced by the large grain, the tool will usually show a longer life. This is based on the assumption that the tools are given a single quench to produce the increase in grain size, which also means that the time or temperature is increased as the grain size is increased. It would appear that the increase of "hot hardness" with increase in grain size in specimens which were given a single quench is also dependent to a great extent upon the fact that either a higher quenching temperature or longer time was necessary to increase the grain size, thereby resulting in a greater solution of the segregate and a greater percentage of austenite in the material as quenched. When the large grain was produced by several quenchings from the same temperature, given an anneal and then quenched, the hardness of the specimen was not nearly as great as when the large grain was produced by a single quench.

(C) Effect of Carbide Segregate on Grain Size

As has been shown, grain size in hardened high speed steel noticeably affects some of the fundamental physical characteristics. It has been questioned whether or not the grain size developed on hardening could be a characteristic of the heat of steel from which it was made. Because of this, bars of 1 inch in diameter were made from the same size ingots, cast from heats of almost exactly the same chemical composition but melted using aluminum, aluminum-silicon, silicon, silicon-zirconium and calcium silicide for deoxidizers. Other variations in melting practice included the use of single and double slags and varying proportions of scrap, wrought iron and ferro-alloys. Samples from these bars were heat treated at the same time at temperatures of 2325, 2350 and 2375 degrees Fahr. (1270, 1285 and 1300 degrees Cent., respectively). A careful measurement of the grain size failed to disclose any material difference.

It has been noticed in treating high speed steel tools that the

grain size in tools of small sections (about 3/8 inch in diameter or thickness and smaller) did not apparently grow as rapidly as when treating larger tools using the same temperature and time, with due allowance for heating rate. The segregate near the outside of a bar is small and well distributed and becomes progressively larger and less well distributed toward the center. These characteristics are more pronounced as the bar increases in size, but an examination of the grain size across the section of a hardened specimen taken from a bar 4 inches or 5 inches in diameter may not disclose a particularly large difference in grain size between the outside and inside of the specimen. This may be explained by the fact that in hardening specimens they are held in the high heat of some 2350 degrees Fahr. (1285 degrees Cent.) for only a few minutes and there is an appreciable temperature gradient between the outside and inside of the specimen at the time of quenching.

In order to obtain samples having the segregate distributed in particles of about the same size entirely across the section, pieces were cut from a forged billet of high speed steel eight inches square, one sample taken 1 inch from the outside, the second halfway between the center and outside, and the third from the center. The three specimens were treated at the same time by quenching from a temperature of 2350 degrees Fahr. (1285 degrees Cent.). Photomicrographs Figs. 29, 30 and 31 taken at the same magnification, 500 diameters, show respectively the grain in the specimens taken 1 inch from the outside, halfway to the center, and near the center of the billet. Each photomicrograph shows definitely that the grain size is the smallest in those areas which have the most uniform and finely divided segregate. Photomicrograph Fig. 30, taken of the specimen halfway to the center, is an excellent example of the restraining influence of the segregate on grain growth, since in the areas having the finely disseminated segregate the grain is considerably smaller than in the surrounding areas having only a few large particles. The restraining effect on grain growth resulting from the finely divided segregate shown in Figs. 29, 30 and 31 is good evidence that the segregate mechanically obstructs grain growth.

In the heating of highly alloyed steels the rate of solution of the segregate in the austenite may affect the rapidity of grain growth, and thus the primary cause of differential grain growth in the samples under discussion results from different amounts of carbide segregate dissolving in solid solution in those areas.

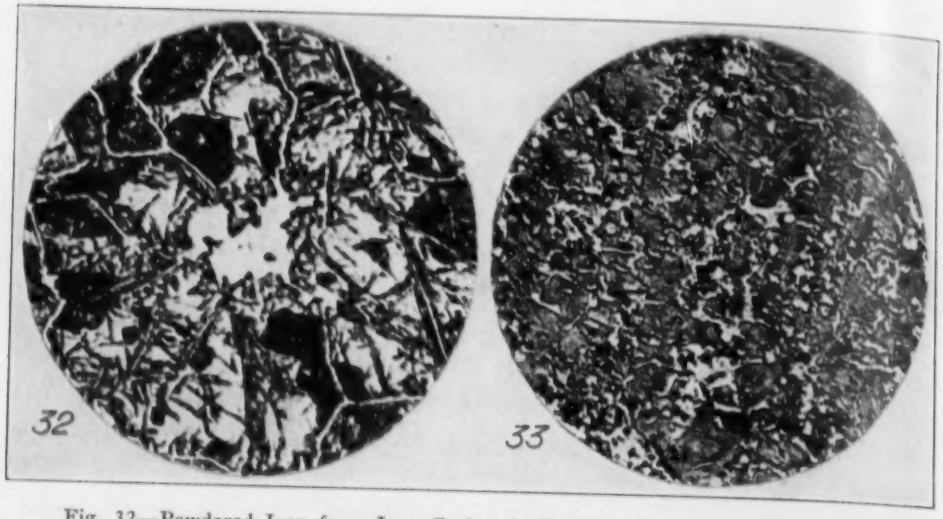
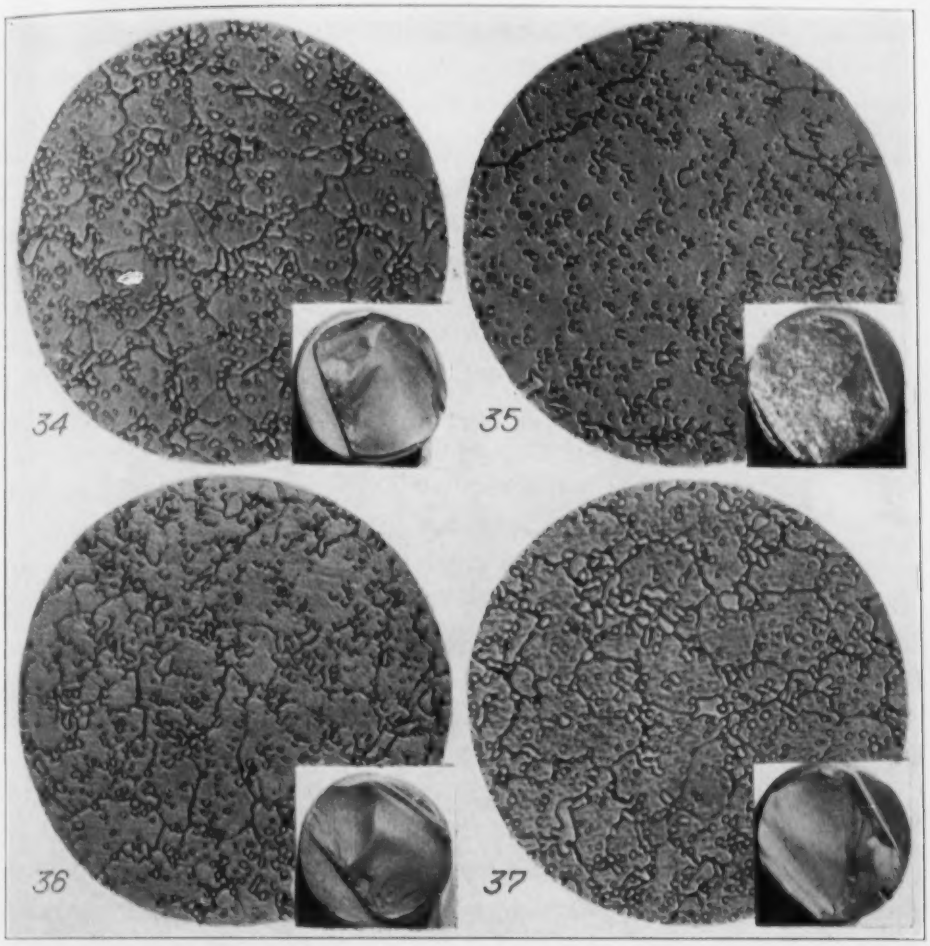


Fig. 32—Powdered Iron from Iron Carbonyl, Carburized from 1.30 Per Cent Carbon, Sintered, Forged and Quenched in Water from 1800 Degrees Fahr. 2 Per Cent Nital. × 500.

Fig. 33—Same as Fig. 32, Mixed with 10 Per Cent TaC Sintered, Forged and Quenched in Water from 1800 Degrees Fahr. 2 Per Cent Nital. × 500.

In order to have better proof that the carbide segregate has a restraining influence on grain growth apart from any grain growth restraint induced by solid solution formation, some very finely powdered iron, 200 mesh and finer, prepared from iron carbonyl was carburized to contain 1.30 per cent carbon. The powder was thoroughly mixed and divided into two parts. To one part was added 10 per cent by weight of tantalum carbide, 325 mesh and finer, and intimately mixed with the carburized iron powder. The two lots of material were then pressed, sintered and forged into ⅝-inch square bars. (Tantalum carbide was chosen because of its high stability and also because tantalum has a greater affinity for carbon than iron and its carbide will not decompose in iron to form tantalum and iron carbide.) After forging, the bars were annealed and then hardened by quenching in water from 1450, 1600, 1700 and 1800 degrees Fahr. (785, 870, 925 and 980 degrees Cent. respectively). All the specimens showed a Rockwell hardness from C 65 to C 66. The two specimens with and without tantalum carbide quenched from 1450 degrees Fahr. (785 degrees Cent.) showed the same grain size, but the specimens with the tantalum carbide showed a grain size of 8½ (Shepherd fracture standard) when quenched from 1700 degrees Fahr. (925 degrees Cent.) and 8¼ when quenched from 1800 degrees Fahr. (980 degrees Cent.), while the specimen without the tantalum carbide showed grain sizes



Figs. 34-37—Fractures ¾ Size of 0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Etched in 4 Per Cent Nital. Not Tempered. × 500.
Fig. 34—Quenched 2350 Degrees Fahr. Fig. 35—Quenched 2350 Degrees Fahr. Requenched 2350 Degrees Fahr. Fig. 36—Quenched 2350 Degrees Fahr. Stress Relieving Treatment. Quenched 2350 Degrees Fahr. Fig. 37—Quenched 2350 Degrees Fahr., Stress Relieving Treatment, Quenched 2350 Degrees Fahr., Stress Relieving Treatment, Third and Final Quench 2350 Degrees Fahr.

of 6½ and 4¾ when quenched from the same temperatures. Photomicrographs Figs. 32 and 33 show the difference in structure between the specimens quenched from 1800 degrees Fahr.

The specimen containing tantalum carbide which had been quenched from 1800 degrees Fahr. (980 degrees Cent.) was then slowly dissolved in dilute hydrochloric acid and the residue obtained. This residue represented 9.34 per cent of the weight dissolved and analyzed 89.78 per cent tantalum and 6.14 per cent carbon, with a negligible amount of tantalum in the filtrate. This is positive proof that the tantalum carbide had not been dissolved in solid solution during hardening.

This large difference in grain size seems a most conclusive bit of evidence upon the mechanical restraining effect on grain growth of carbide particles imbedded in the material.

(D) Grain Size as Affected by Repeated Hardening

It is well known that the re-hardening of high speed steel, when performed under certain conditions, results in the steel having a so-called "fish scale" or "marble" fracture. Brophy and Harrington (7) (8), in several papers on the subject, have apparently established that the abnormal grain growth which causes such a fracture results from straining, and that there is a direct relationship between the size of the grain and the degree of stress existing in the steel at the time of reheating for hardening. They also pointed out that when the material is highly stressed such stresses must be removed by annealing for extended periods of time to prevent abnormal grain growth.

In further investigation of this subject it was found that annealing at the usual temperatures of from 1550 to 1650 degrees Fahr. (840 to 900 degrees Cent.) may or may not prevent abnormal grain growth on subsequent hardening. It was found that the controlling factor was probably the rate of heating through the recrystallization range of 1375 to 1440 degrees Fahr. (750 to 780 degrees Cent.) and that consequently annealing from 1550 to 1650 degrees Fahr. (840 to 900 degrees Cent.) may or may not prevent abnormal grain growth dependent on the rate of heating.

It is apparent that the rate of heating must be of considerable influence, since a double treatment using a preheat of 1550 degrees Fahr. (840 degrees Cent.) for 1 hour and a high heat of 2350 degrees Fahr. (1285 degrees Cent.) results in abnormal grain growth and a "fish scale" fracture in a specimen 1 inch in diameter. The preheat for the second hardening of 1550 degrees Fahr. (840 degrees Cent.), even though held at temperature as long as three hours, does not prevent the abnormal grain growth, which would indicate that the temperature was above the recrystallization range.

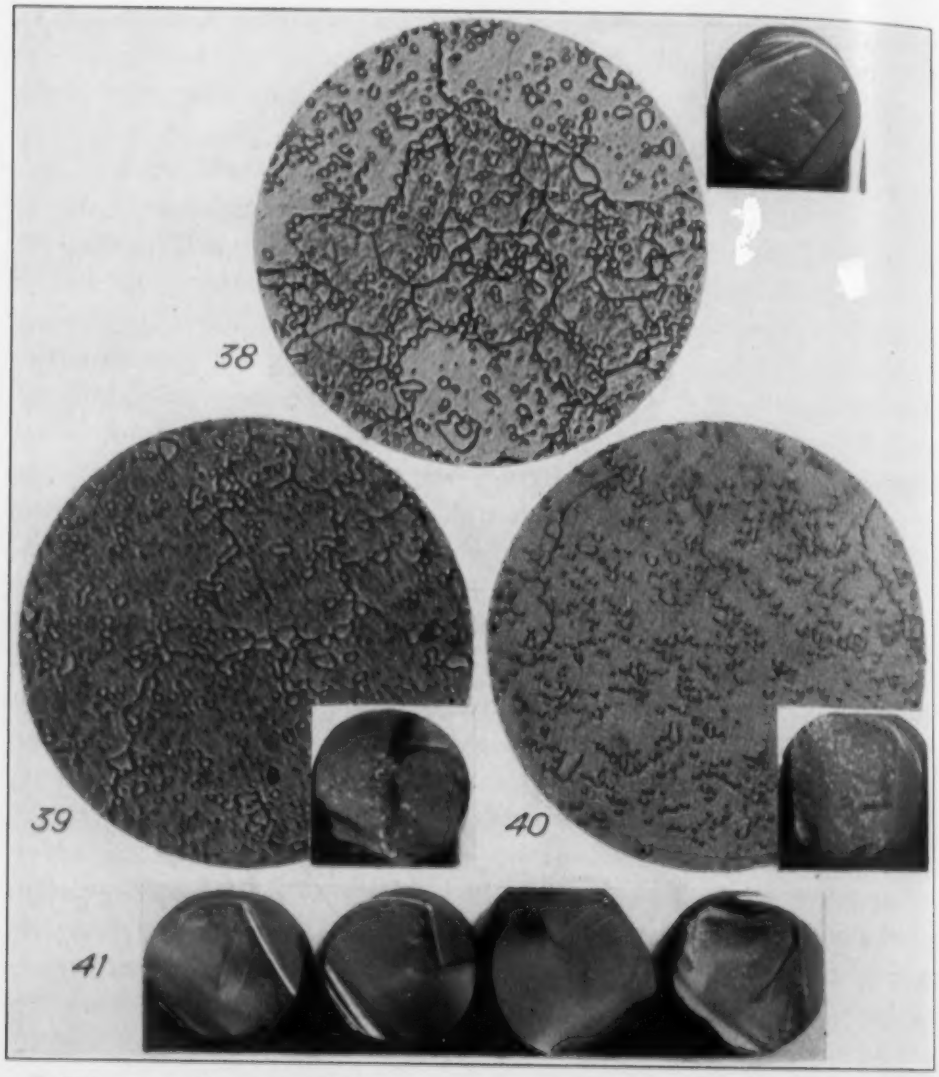
A number of annealed specimens of 0.70 per cent carbon, 18-4-1 high speed steel 1 inch in diameter, numbered from 26 to 36, were selected and given an original hardening treatment, consisting of a preheat to 1550 degrees Fahr. (840 degrees Cent.) for 1 hour and a high heat to 2350 degrees Fahr. (1285 degrees Cent.). The specimens were oil-quenched to room temperature, and then given vari-

ous treatments. The specimens as originally treated had a fracture and grain size as shown in Fig. 34. Specimen 27 was then retreated in exactly the same manner, using the same time and temperature and the resultant structure is shown in Fig. 35. The grain is very large and the fracture of the specimen shows a characteristic "fish scale" or "marble" appearance. Specimen 28, having already a single quench, was given a stress relief treatment by heating slowly to 1100 degrees Fahr. (600 degrees Cent.), transferring to a furnace at 1375 degrees Fahr. (745 degrees Cent.), and allowed to remain at this temperature until uniformly heated. The temperature was then advanced by 25-degree increments and the specimen held at each temperature for one-half hour until a temperature of 1450 degrees Fahr. (785 degrees Cent.) was reached when the specimen was transferred to the preheat furnace at 1550 degrees Fahr. (840 degrees Cent.) before transferring to the high heat furnace for the final heating. Approximately three hours' time was required to heat the specimen from room temperature to 1550 degrees Fahr. (840 degrees Cent.). The specimen was then quenched from 2350 degrees Fahr. (1285 degrees Cent.) and the structure and fracture are shown in Fig. 36. It will be noticed that the grain size as indicated by both the fracture and micrograph is substantially the same as when a single quench was used.

Specimen 29 after the single quench was given a stress relief treatment as described, re-quenched, given a second stress relief treatment, and then given a third quench. The grain and fracture remained fine and compared in all respects to the specimen which had only been given a single quench (see Fig. 37). Specimen 30 was given the same procedure as specimen 29 until it was quenched the fourth time, but this specimen shows slight "fish scale" fracture in the center and a mixed grain size (see Fig. 38). Specimen 31 was quenched and alternated with the stress relief treatment until it was quenched the fifth time. It showed (Fig. 38) a typical "fish scale" fracture and structure having large and small grains mixed.

A specimen (Fig. 35) which had been given a double quenching and showed abnormal grain growth was then subjected to a number of treatments to reduce grain size, but regardless of treatment there was no indication that this could be accomplished (Fig. 40).

Assuming that stresses are the primary cause of the abnormal grain growth, then apparently our procedure of stress removal did not quite remove all of the stresses in specimens of this size. The



Figs. 38-41—Fractures ¾ Size of 0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Specimens Quenched in 4 Per Cent Nital. × 500.

Fig. 38—Alternately Quenched from 2350 Degrees Fahr. and Given Stress Relieving Treatment Until Quenched Fourth Time from 2350 Degrees Fahr. Not Tempered. Fig. 39—Alternately Quenched from 2350 Degrees Fahr. and Given Stress Relieving Treatment Until Quenched Fifth Time. Fig. 40—Quenched from 2350 Degrees Fahr., Requenched 2350 Degrees Fahr. Given Stress Relieving and Other Heat Treatments to Reduce Grain Size. Fig. 41—Specimens Quenched Alternately from 2350 Degrees Fahr. and Given Stress Relieving Treatment. Specimens from Left to Right Quenched 3, 4, 5, and 6 Times Respectively.

following stress relief treatment was then applied: The specimen was heated slowly to 1100 degrees Fahr. (600 degrees Cent.), transferred to a furnace at 1200 degrees Fahr. (650 degrees Cent.), and allowed to attain this temperature. It was then heated to 1325 degrees Fahr. (715 degrees Cent.), held at this temperature for one-half hour and then the temperature increased by 25-degree in-

crements until 1375 degrees Fahr. (750 degrees Cent.) was attained, where it was held for one-half hour. From 1375 to 1450 degrees Fahr. the temperature was increased by 25-degree increments, holding at each temperature 1 hour, after which the specimen was heated to 1550 degrees Fahr. (840 degrees Cent.) for preheating preparatory to transferring to the high temperature furnace for hardening. The total time required was about seven hours. The specimens were then alternately quenched from 2350 degrees Fahr. (1285 degrees Cent.) and given this stress relief treatment.

Specimens 33, 34 and 35 quenched the third, fourth and fifth times showed no indication of abnormal grain growth, either by fracture or by micrograph, but specimen 36 quenched the sixth time did show a slight indication of abnormal grain growth. (Fig. 41). The grain size was substantially the same in all of the specimens, comparing in size with the sample which had been given a single treatment. The fractures of the specimens are shown in Fig. 41.

These simple experiments would seem to confirm the deductions of Brophy and Harrington in that abnormal grain growth is caused by stresses and they also give evidence as to the length of time and temperatures necessary to remove completely such stresses. These specimens show clearly that stresses are much greater near the center of the specimen than toward the outside and that it would undoubtedly be more difficult to remove such stresses with an increase in the size of the specimens. It is also likely that the temperature and rate of grain germination may be influenced somewhat by the degree of stress existing at the time the specimen is heated.

(E) Grain Size as Affected by Furnace Atmosphere in Hardening

It is important, in the hardening of fine edged tools, to heat them in such a manner that the fine cutting edges will not be damaged. For many years it has been known that furnace atmospheres materially affect the surface of high speed steel when heating for hardening. Furnaces with controlled atmospheres have been developed primarily to produce a better surface on hardened tools. Shortly after this it was noted that in certain atmospheres the steel could be held much longer at the hardening temperature without material grain growth. Tour (9) and Phillips and Weldon (10) published papers on this phenomenon. Their deductions were prac-

tically the same, namely, that the gases surrounding the steel during heating for hardening did in fact materially affect the grain size as well as the temperature at which the carbide segregate would fuse.

About four years ago the speaker, on investigating this puzzling phenomenon, at first came to the same conclusions as the investigators mentioned, but when attempting to check his results in a different furnace and with gases synthetically mixed, different results were obtained. Further investigations led to the conclusions that the gases surrounding the steel were materially affecting the temperature of the steel as well as the time of heating. A number of investigators have taken exception to these conclusions and for this reason experiments were again made to show that by varying the composition of the gas, a change could be made in the rate of heating.

The first group of samples were treated in a controlled atmosphere furnace. The samples were taken from the same bar of 0.70 per cent carbon, 18-4-1 high speed steel. The samples were machined by removing one-eighth inch of surface to give a cylindrical bar 2 inches in diameter by 2 inches long; a $\frac{1}{32}$-inch diameter hole was drilled longitudinally in the center to a depth of $1\frac{1}{4}$ inch. Three thermocouples were used: one inside the specimen, another flat against the surface of the specimen; the third was the furnace control couple located about 6 inches above the hearth. The temperature of the hearth was taken immediately before placing the specimen in the furnace and immediately after its removal. New thermocouples were used for each specimen, and temperature readings were taken every 30 seconds. All the specimens were preheated alike in the same furnace and for the same period of time, which was 35 minutes at 1550 degrees Fahr. (840 degrees Cent.).

For the first two specimens (Nos. 41 and 42) the temperature control was set at 2360 degrees Fahr. (1280 degrees Cent.). An atmosphere containing 4.6 per cent CO_2 and 11.4 per cent CO was used for specimen 41 and an atmosphere of 11.0 per cent CO_2 and 1.0 per cent CO was used for specimen 42. The results are given in Fig. 42, which definitely shows that the specimen heated in the gas containing the greater percentage of CO_2 rose in temperature more rapidly and attained a higher temperature. It is interesting to note that specimen 41 never reached the hearth temperature, while specimen 42 did.

From this diagram it appeared that possibly specimen 41 had not been held a sufficient length of time to reach the hearth temperature, so specimen 43 was heated in the same atmosphere as specimen 41 and held for 20 minutes with the results shown in Fig. 43; again the specimen did not become as hot as the hearth.

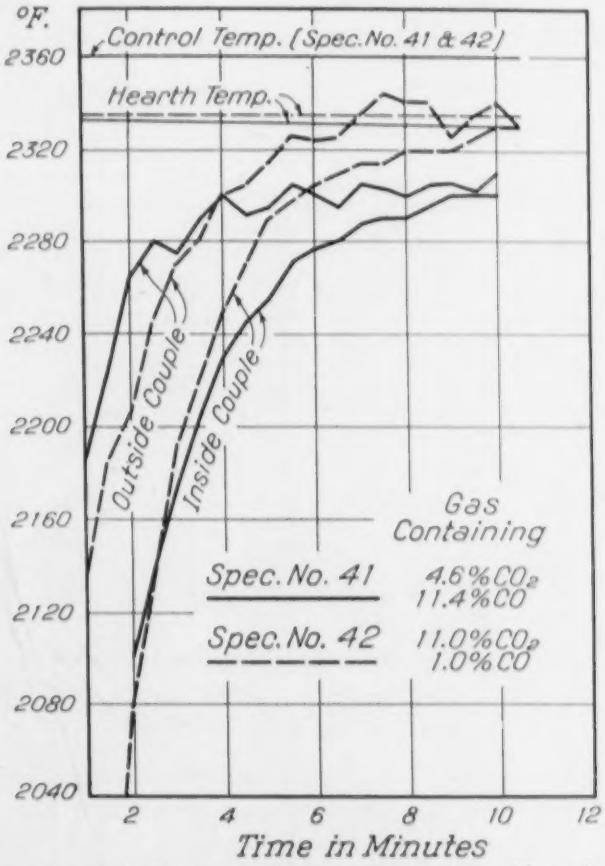


Fig. 42—Diagram Showing Temperature Rise of Specimens Subjected to Varying Percentages of CO and CO_2.

By experiment a temperature was selected so that specimen 44 could be heated at the same rate and to the same temperature in a gas containing 3.8 per cent CO_2 and 11.6 per cent CO as specimen 42, which was heated in gas containing 11.0 per cent CO_2 and 1.0 per cent CO. It was found that the temperature control should be set at 2390 degrees Fahr. (1305 degrees Cent.) to obtain this heating rate.

Fig. 44 shows a comparison of the temperature reading for specimen 42 compared with specimen 44. Photomicrographs shown in Figs. 45 and 46 taken halfway between center and outside of these

specimens show no noticeable difference in grain size and the hardness of the two specimens was exactly the same, namely, Rockwell 66 C. Similar experiments were made using the same furnace with about the same atmosphere but varying the temperature and in every instance the results were similar to those given.

It must not be inferred from these experiments that the amount of CO and CO_2 gases present alone determines the rate of heating.

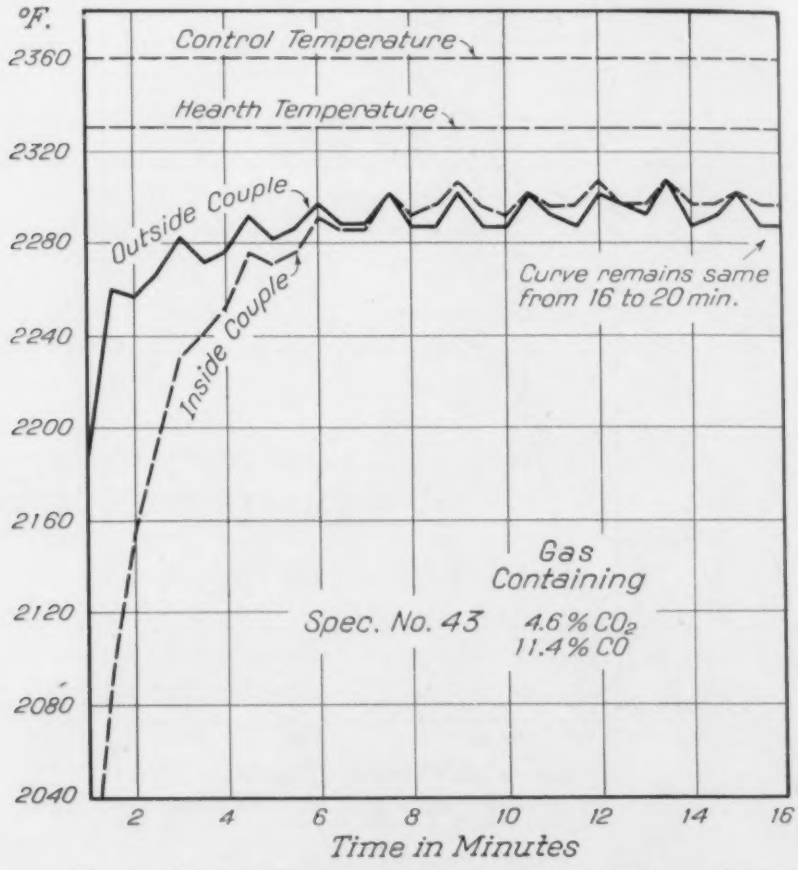


Fig. 43—Diagram Showing Temperature Rise of Specimens When Subjected to Furnace Atmospheres Indicated in Diagram.

It is possible to keep the amount of CO and CO_2 gases constant and yet vary the rate of heating by a change in the composition of the remainder of the gas.

If atmospheres do affect grain size then it should be noticeable with radically different atmospheres; to test this, additional experiments were made with an entirely different experimental arrangement. A special quartz tube was used for the heating chamber and specimens were made from the same high speed steel bars,

1 inch diameter, machined to ¾ inch diameter by 1¼ inch long with a ¼-inch diameter hole drilled longitudinally in the center to a depth of ¾ inch. Three platinum, platinum-rhodium thermocouples were used, one inside the specimen, one against the outside of the specimen and the third for the chamber temperature (which was taken immediately before the specimen was inserted and immediately after the specimen was removed, at the same location as

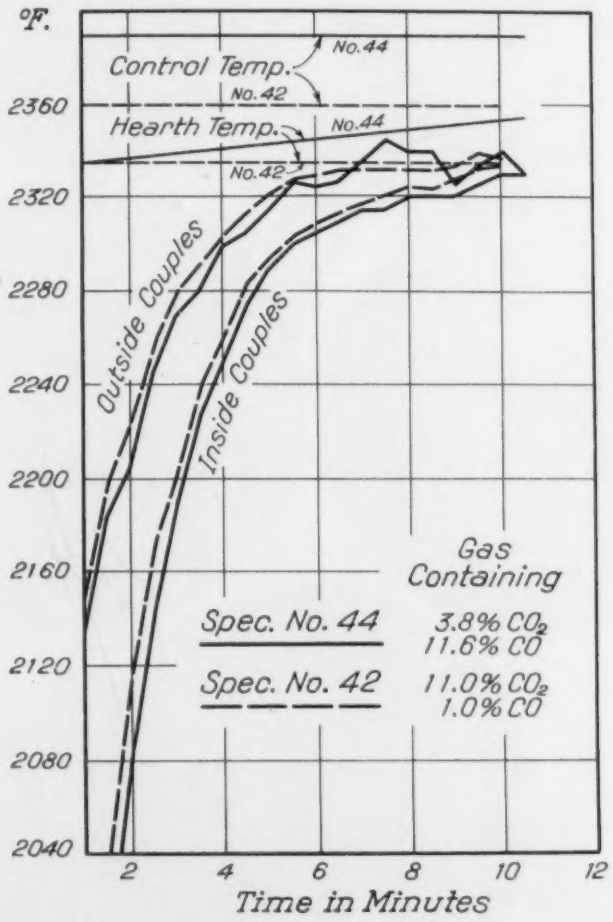
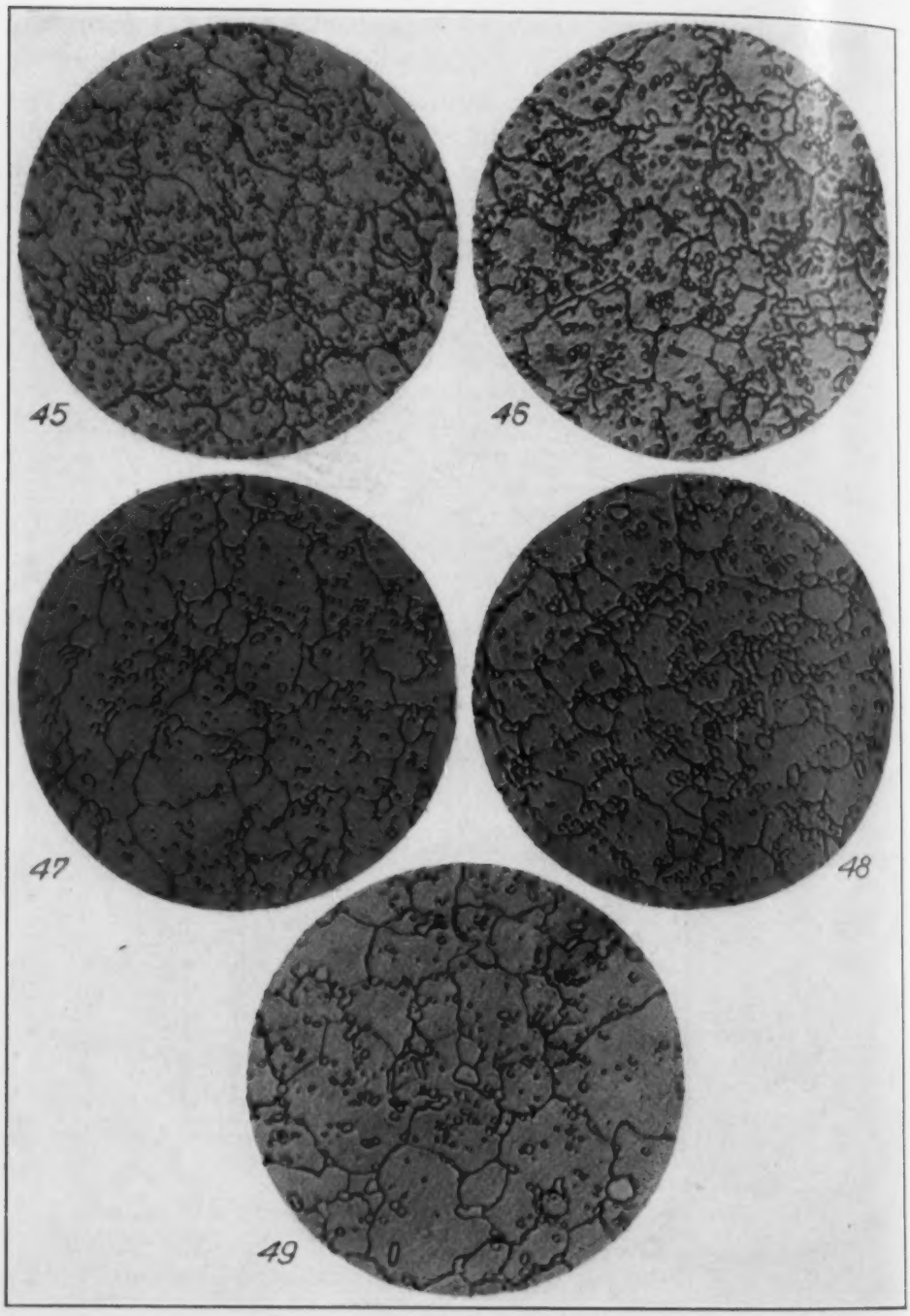


Fig. 44—Diagram Showing Specimens Nos. 42 and 44 When Subjected to Furnace Atmospheres Indicated in Diagram.

the specimen). The thermocouples were calibrated and renewed when necessary for each specimen. Temperatures were recorded every 30 seconds.

Specimens without preheating were heated in pure CO, pure CO_2, and air, and during the heating the gases were flowing slowly



Figs. 45-49—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Etched in 4 Per Cent Nital. × 500.

Fig. 45—Quenched from Temperature Shown in Fig. 42. Not Tempered. Fig. 46—Quenched from Temperature Shown in Fig. 44. Not Tempered. Fig. 47—Quenched from Temperature Shown in Fig. 50. Not Tempered. Fig. 48—Quenched from Temperature Shown in Fig. 50. Not Tempered. Fig. 49—Quenched from Temperature Shown in Fig. 50. Not Tempered.

through the tube, each at the same rate. The tube was sufficiently long to preheat the gases and tests were made to be certain that the gases were at the correct temperature at the position of the specimen. The tube temperature was originally set for 2340 degrees Fahr. (1280 degrees Cent.) and the specimens were heated for various lengths of time in the respective gases. By comparing the

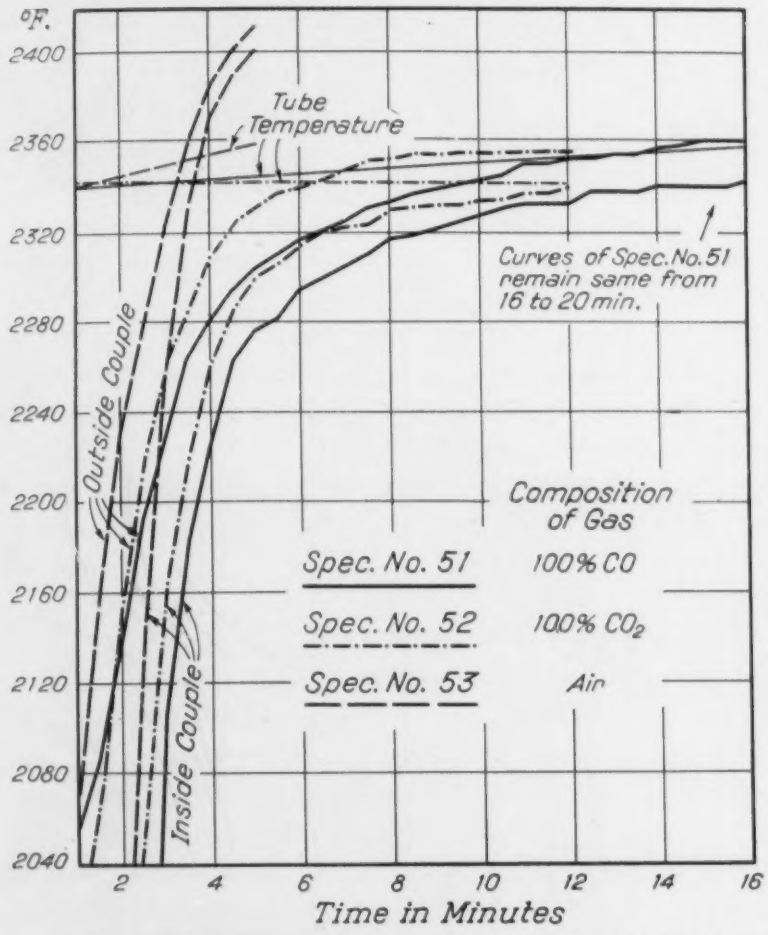
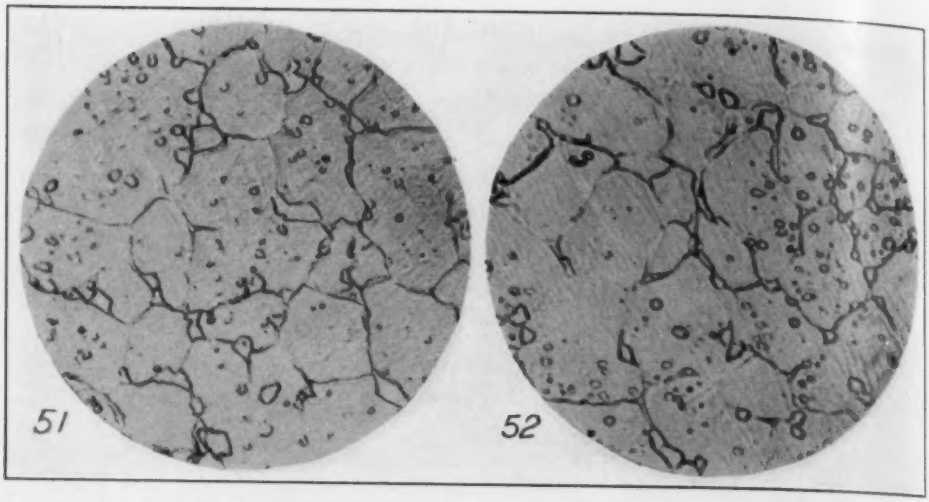


Fig. 50—Time-Temperature Curves for Specimens 51, 52 and 53 with Different Gas Atmospheres as Shown in Diagram.

grain size of the different specimens 51, 52 and 53 (Figs. 47, 48 and 49) it was found that specimen 51, held for 20 minutes in CO gas, was similar in grain size to specimen 52 held 12 minutes in CO_2 gas, and to specimen 53 held five minutes in air, but when the time-temperature curves of the specimens shown in Fig. 50 are compared the reason is evident. The inside of specimen 51 heated in CO gas did not reach the tube temperature in 20 minutes. The inside of specimen



Figs. 51 and 52—0.70 Per Cent Carbon. 18-4-1 High Speed Steel. Quenched from Temperature Shown in Fig. 53. Not Tempered. Etched in 4 Per Cent Nital. × 500.

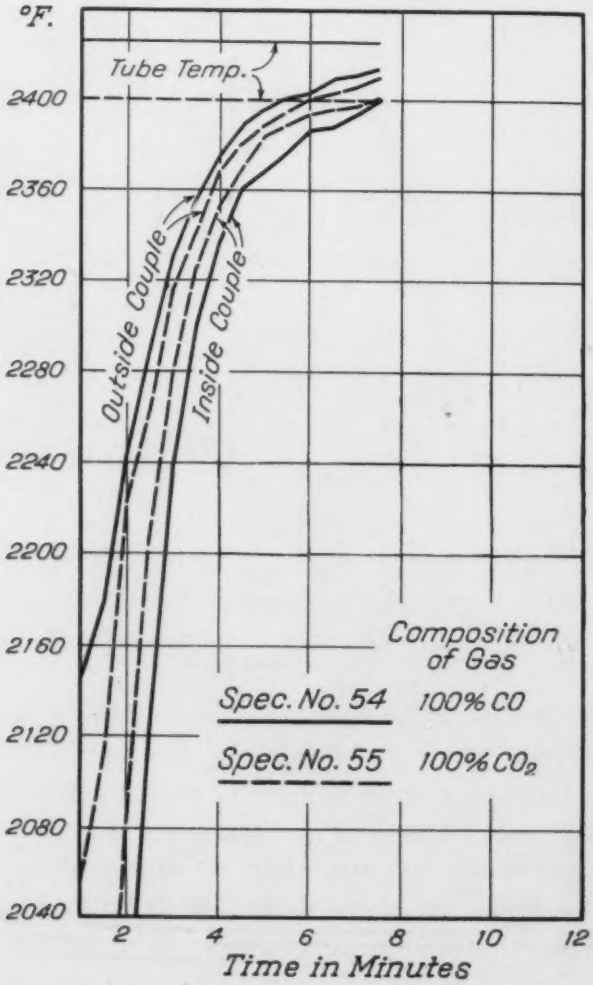


Fig. 53—Time-Temperature Curves for Specimens 54 and 55 in Different Gas Atmospheres as Shown.

52 heated in CO_2 gas reached the tube temperature in 12 minutes and the surface of the specimen exceeded the tube temperature, while the specimen heated in air far exceeded the tube temperature both on the inside and outside. This specimen plainly shows self-heating or heating from oxidation.

The temperature and time were adjusted so that the carbide segregate would just begin to fuse using pure CO gas; this was found to be 2425 degrees Fahr. (1325 degrees Cent.) for 7½ minutes for specimen 54; the microstructure is shown in Fig. 51. The time-temperature curve of specimen 54 was then reproduced for specimen 55 but pure CO_2 gas was used. It was found that a tube temperature of 2400 degrees Fahr. (1310 degrees Cent.) reproduced about the same curve, using CO_2 gas. Fig. 52 is a photomicrograph of specimen 55, showing a structure quite comparable to specimen 54, heated in CO gas. The time-temperature curves of specimens 54 and 55 are shown in Fig. 53.

The experiments using pure gas were made at other temperatures and similar results were obtained.

I feel that the described investigations into the behavior of high speed steel are very incomplete, yet some of these experiments should have practical applications and offer suggestions for further study.

Summary

1. The carbide segregate in an 18-4-1 high speed steel is not materially changed in composition by heat treatment, by distribution or by outward form.

2. The outward appearance of the carbide segregate seems to influence the rate of diffusion and the hardenability.

3. The strength and plasticity of high speed steel is materially affected by the grain size, by the distribution of the segregate and by the outward form of the segregate.

4. The thermal conductivity is not materially affected by the grain size or by the distribution and form of the segregate.

5. The hardness at elevated temperatures is materially affected by grain size and by the outward appearance of the segregate. The distribution of the segregate is of inferior influence.

6. The susceptibility to grain growth is not a characteristic of the method used in melting and deoxidizing the steel.

7. The carbide segregate which is not dissolved in solid solution in the steel mechanically obstructs grain growth.

8. The stresses which result from the initial hardening of high speed steel require a considerable length of time for removal at comparatively low temperatures. The stresses when fully removed will only slightly affect the physical properties of the steel on re-hardening.

9. The grain size of the steel is affected by the composition of the furnace atmosphere surrounding the steel during hardening, only as the composition of the gases affect the rate of heating and the temperature of the specimen.

ACKNOWLEDGMENTS

A number of investigators have given their help in the preparation of the material for this lecture and I acknowledge to these due credit and give their names in alphabetical order.

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Mr. R. S. Rose

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ADDENDA

NATURE AND CHARACTERISTICS OF CARBIDE SEGREGATE

In Tables VII and VIII will be found weights and percentages of samples and residues given in Tables I and III of the lecture.

Table VII

Specimen Number	Weight of Sample Grams	Weight of Residue Grams	Weight Dissolved in Filtrate Grams
1	132.2732	36.1084	96.1648
2	141.1880	36.7960	104.3920
3	139.1682	39.3430	99.8252
4	86.0604	23.2046	62.8558
5	108.7808	28.2720	80.5088
6	67.4398	17.5020	49.9378
9X	96.7362	35.1550	61.5812
10X	116.3798	32.7650	83.6148
11X	117.5784	31.2900	86.2884
12X	138.7732	45.3540	93.4192

Table VIII

Specimen Number	Residue Per Cent	Amount Dissolved in Filtrate Per Cent
1	27.29	72.70
2	28.27	71.72
3	32.68	67.31
4	26.06	73.93
5	25.98	74.01
6	25.95	74.04
9X	26.96	73.03
10X	28.15	71.84
11X	26.61	73.38
12X	36.34	63.65

Table IX

Specimen Number	Tungsten	Chromium	Vanadium
1	0.1055	4.1262	0.1102
2	1.4980	4.3338	0.6628
3	1.2632	4.8332	0.5088
4	0.6341	4.4991	0.9482
5	0.7533	4.6454	0.5093
6	0.5407	4.1732	0.2523
9X	0.1674	2.0785	0.6073
10X	0.2085	4.0585	0.8539
11X	0.5606	4.4970	0.7556
12X	0.1043	3.5412	0.1712

In Table IX is given the analysis of the filtrates of the samples used in Tables I and III of the lecture.

X-Ray Diffraction Data

Focusing cameras of the Phragmen type were used. Chromium radiation was used throughout the series. Residue No. 12X was also studied with iron and copper radiation and the lattice constants from these independent determinations with the three different wave lengths serve nicely to test the absolute accuracy. The results were a_0 (Cr radiation) 11.048, (Fe radiation) 11.048, (Cu radiation) 11.044, showing an error of about 0.02 per cent; however, the accuracy of the lattice dimensions relative to one another somewhat exceeds this.

The tables list the relative intensity of the lines; the value of $\sin^2\theta$; the sum of the squares of the indices of the reflecting planes $(h^2+k^2+l^2)$; the radiation of the CrK series which produced the lines (α_1, α_2 and β); and finally the lattice constant a_0, which is calculated from the formula for cubic crystals:

$$\frac{\sin^2\theta}{h^2+k^2+l^2} = \frac{\lambda^2}{4a_0^2}$$

The carbide phase accounting for most of the lines is face-centered cubic in structure with a_0 varying from 11.028 to 11.053 A.U. The size of the unit cell, the X-ray lines obtained from this phase and the relative intensities of the lines are all in agreement with those tabulated by Westgren and his associates for the principal carbide phase in high speed steels.

A new set of lines appears on photograms of the residue from the low carbon steels and increases in intensity as the carbon decreases. These lines were weaker and more difficult to measure, but visual comparison of the films shows the same set of lines very faintly in residue No. 11X, stronger in residue No. 10X and strongest in residue No. 9X.

These lines were first compared with the diffraction pattern for a carbide phase found in high speed steel in small amount by Morral, Phragmen and Westgren which they found was of a composition approximately Fe_4C with a little of the iron replaced with W so as to be $(Fe,W)_4C$. It had a face-centered cubic structure of $a_0 =$ 10.51Å analogous to Cr_4C with $a_0 =$ 10.64Å. The lines do not agree with those for a 10.51Å face-centered cubic unit cell, while a

Table X

Residue No. 1
a_0 equals 11.028 Face-centered cubic lattice of "Fe_3W_3C"
Camera No. 3 (outer camera)

Intensity	$Sin^2\theta$	$(h^2+k^2+l^2)$	Radiation CrK Series
W	0.1420	16	β
M	0.1723	16	α
M	0.2041	19	α
M	0.2132	24	β
M	0.2401	27	β
S	0.2548		
S	0.2579	24	α
S	0.2872		
S	0.2899	27	α
M	0.3437	32	α
VW	0.3867	36	α
W	0.4539	51	β
VW	0.4726	44	α
W	0.5250	59	β
M	0.5479	51	α_1
W	0.5971	67	β
M	0.6345	59	α_1
W	0.6353	59	α_2
M	0.6420	72	β
W	0.6677	75	β
W	0.6876	64	α_1
S	0.7187	67	α_1
M	0.7215	67	α_2
VS	0.7725	72	α_1
S	0.7759	72	α_2
S	0.8051	75	α_1
M	0.8078	75	α_2
M	0.8806	99	β
M	0.8897	83	α_1
W	0.8935	83	α_2
VW	0.9004	84	α_1
M	0.9245	104	β
M	0.9427	88	α_1
W	0.9468	88	α_2
M	0.9505	107	β

Table XI

Residue No. 9X CrK Radiation
a_0 equals 11.052

Intensity	$Sin^2\theta$	$(h^2+k^2+l^2)$	Radiation
W	0.2586	24	α
W	0.2678		β
M	0.2785		β
M	0.2910	27	α
M	0.3241		α
M	0.3351		α
W	0.3438	32	α
VW	0.5453	51	α_1
VW	0.6150		β
VW	0.6287		α
W	0.6294	59	α_1
W	0.6378	72	β
VW	0.6536		β
VW	0.6631	75	β
W	0.7155	67	α_1
M	0.7421		α_1
M	0.7441		α_2
M	0.7669	72	α_1
M	0.7709	72	α_2
W	0.7873		α
W	0.8016	75	α_1
W	0.8037	75	α_2
W	0.8603		α_1
W	0.8625		α_2
M	0.9324		α_1
M	0.9348		α_2

Table XII

Residue No. 10X CrK Radiation
a_o equals 11.053

Intensity	$\sin^2\theta$	$(h^2+k^2+l^2)$	Radiation
M	0.5459	51	α_1
VW	0.5940	67	β
W	0.6167		
M	0.6308	59	α_1
M	0.6331	59	α_2
M	0.6389	72	β
VW	0.6637	75	β
VW	0.7018		
M	0.7157	67	α_1
M	0.7173	67	α_2
M	0.7430		
VS	0.7689	72	α_1
VS	0.7717	72	α_2
M	0.7882		α_1
M	0.7904		α_2
M	0.8017	75	α_1
M	0.8034	75	α_2
M	0.8599		α_1
M	0.8627		α_2
W	0.8760	99	β
W	0.8858	83	α_1
W	0.9198		
W	0.9334		
W	0.9475	105	β

Table XIII

Residue No. 11X CrK Radiation
a_o equals 11.051

Intensity	$\sin^2\theta$	$(h^2+k^2+l^2)$	Radiation
M	0.5458	51	α_1
M	0.5476	51	α_2
VW	0.5940	67	β
W	0.6314	59	α_1
W	0.6338	59	α_2
M	0.6390	72	β
VW	0.6654	75	β
M	0.7163	67	α_1
M	0.7182	67	α_2
W	0.7415		α_1
W	0.7436		α_2
VS	0.7697	72	α_1
VS	0.7720	72	α_2
M	0.8012	75	α_1
W	0.8035	75	α_2
VW	0.8775	99	β
VW	0.8870	83	α_1
W	0.8889	83	α_2
W	0.9210		
W	0.9324		α
W	0.9400	88	α_1
W	0.9480	107	β

face-centered cubic cell of a_o equals 12.070 will explain the position of many of the lines, although the $(FeW)_4C$ phase might have been expanded by the large tungsten atoms, one would not expect as great an expansion as this (10.5 to 12.07). Furthermore, some of the strongest "extra" lines which occur at lower angles of reflection (Table XI) were left unexplained. The relative intensities were

Table XIV

"Extra lines" not belonging to Fe_3W_3C. Read on films from residues numbers 9X, 10X and 11X where they appear strongest. Compared with spectrum Fe_2W calculated for Cr radiation.

			Observed "extra lines" CrKα					
			Residue No. 9X		Residue No. 10X		Residue No. 11X	
(hkl)	Intensity	$\sin^2\theta$	Intensity	$\sin^2\theta$	Intensity	$\sin^2\theta$	Intensity	$\sin^2\theta$
102	M	0.166						
110	S	0.233						
103	S	0.276						
200	M	0.311						
112	S	0.322	M	0.3241				
201	S	0.333	M	0.3351				
004	W	0.353						
104	W	0.430						
203	W	0.509						
210	W	0.545						
211	W	0.567						
114	VW	0.586						
105	W	0.627	VW	0.6287				
212	W	0.632						
300	W	0.701			VW	0.7018		
213	S	0.743	M	$\{0.7421a_1$ $\{0.7441a_2$	M	0.7430a	W	$\{0.7415a_1$ $\{0.7436a_2$
302 006	S	0.790	W	0.7873	M	$\{0.7882a_1$ $\{0.7904a_2$		
205	S	0.861	W	$\{0.8603a_1$ $\{0.8625a_2$	M	$\{0.8599a_1$ $\{0.8627a_2$		
106	W	0.869						
214	W	0.898						
220	S	0.935	M	$\{0.9324a_1$ $\{0.9348a_2$	VW	0.9334a	W	0.9324
					Unaccounted for			
					W	0.9205	W	0.9210

not calculated for it appeared very improbable that the extra lines were from this phase.

Westgren and Phragmen have reported another Fe-W-C phase, that appears more abundantly in melts of lower carbon content and seems to be richer in tungsten and lower in carbon than the high speed steel carbide Fe_3W_3C. It appeared to be rhombohedral with a complex lattice and to be an intermediate FeW phase which had been found in ferrotungsten of about 50 per cent tungsten by Arnfelt. This structure was solved by Arnfelt and Westgren. *(Jernkontortels Annaler,* 1935, 185). Its formula ordinarily given is Fe_3W_2 (analogous to Fe_3Mo_2), approximates more closely Fe_7W_6. The rhombohedral unit cell contains 13 atoms. Neither the intensities nor positions of the extra lines of the present films are in agreement with the reflections listed for this structure.

Finally, the lines were successfully explained by a phase not expected from the constitution diagram suggested by Grossmann and Bain, the phase Fe_2W. The structure was derived by Arnfelt and Westgren. *(Jernkontortels Annaler,* 1935, pages 185, 196.) The unit cell contains 12 atoms and is hexagonal, D^4_{6h}.

In Table XIV are listed the reflections for Fe_2W with $\sin^2\theta$

Table XV

Residue Number	a_0
1	11.028
2	11.035
3	11.035
4	11.04
5	11.035
6	11.035
9X	11.052
10X	11.053
11X	11.051
12X	11.048

converted over to the values for CrK radiation. It will be seen that the lines from carbide residues 10X, 11X and 9X agree satisfactorily in relative intensity as well as in position ($\sin^2\theta$) with these Fe_2W lines. Only one weak line remains unexplained.

Melting Point of Fe_3W_3C

Material prepared in proportions approximately to conform to the composition of Fe_3W_3C had a melting point of 3190 ± 10 degrees Fahr. Analysis of this material after determination of the melting point was carbon 1.49 per cent, tungsten 75.23 per cent, iron 22.01 per cent.

Physical Properties

Torsion Tests

The torsion tests were made on a machine especially built by Emery-Baldwin-Southwark, based on the machine described by Emmons in the paper cited. All specimens were tested in triplicate and the average of the three specimens was plotted. In every instance the three specimens gave results sufficiently consistent with one another that it was unnecessary to repeat any of the tests.

Hardness at Elevated Temperatures

The equipment used for making these tests has been fully described by Harder and Grove in the article cited. Fig. 54 shows the testing equipment completely assembled and Fig. 55 shows a detail of the testing assembly. The furnace and compression units were first heated to 1200 degrees Fahr. The compression assembly was then lowered, the cylinders to be tested inserted, and the unit raised to the position shown in Fig. 55. From 5 to 7 minutes were

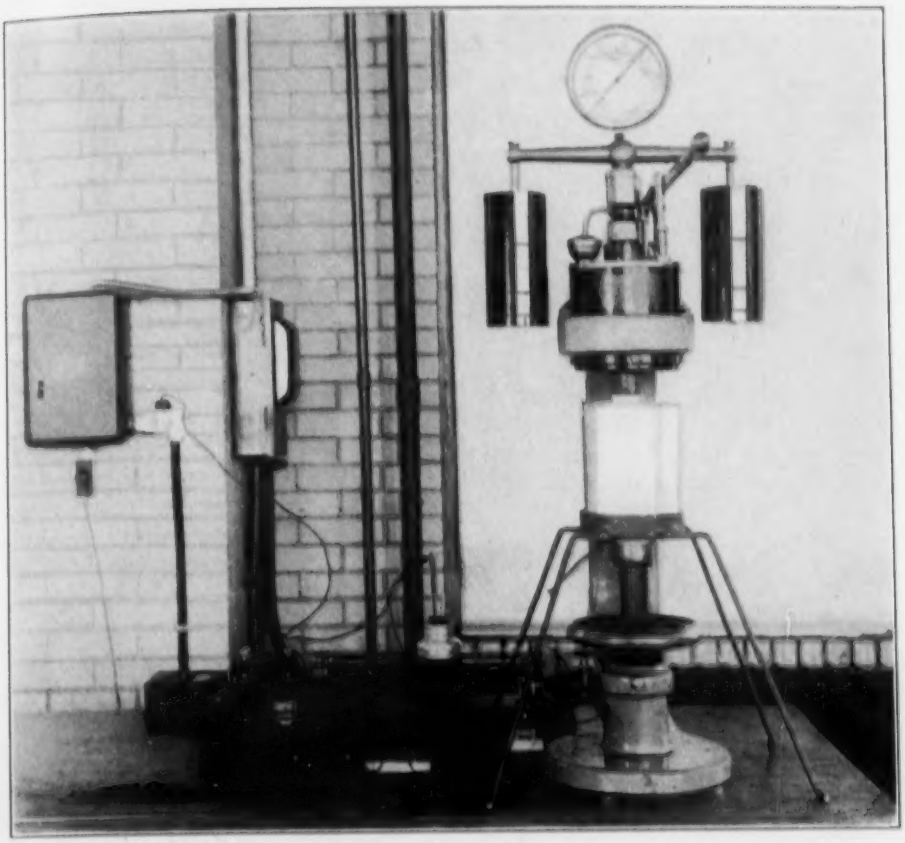

Fig. 54—Shows Testing Equipment Completely Assembled.

required for the specimens to reach a temperature of 1200 degrees Fahr. In each test the specimens were maintained at the testing temperature of 1200 degrees Fahr. for 20 minutes after reaching that temperature. A load of 3000 kilograms was then applied for 30 seconds.

The specimens were then removed and the width of the impression measured with a Brinell microscope. The width in millimeters times the length (10 millimeters) gave the area of the indented surface. The cylinder hardness was calculated from the following equation:

$$\text{Cylinder Hardness} = \frac{\text{Applied Load in Kilograms}}{\text{Area of Indentation in Square Millimeters}}$$

The Brinell hardness was then determined by multiplying the cylinder hardness by 1.52. The work of Harder and Grove demonstrated the dependability of this factor and procedure. Cowdrey's

Table XVI

Hardness of specimens after being subject to Mutual Indentation Hardness Tests at 1200 degrees Fahr.
Size of test specimens: 1 centimeter diameter by 1 centimeter long.

Specimen Number	Mutual Indentation Hardness—Brinell	Hardness Sample A Rockwell C	Hardness Sample B Rockwell C
13	337	59.5	59.8
13	345	59.8	59.5
13-T	351	58.1	58.1
13-T	326	56.5	56.4
14	380	61.7	61.9
14	390	61.7	62.8
14-T	388	61.5	61.6
14-T	404	61.7	61.7
15	404	63.3	63.4
15	407	63	63
15-T	415	63.4	62.8
15-T	415	63.4	63
16	386	61.4	61.3
16	383	61	61
16-T	369	61.2	61.1
16-T	380	61.2	61.4
17	380	60.7	60.7
17	390	60.9	61.1
17-T	351	60.2	60.2
17-T	358	60	60.1
18	383	61.5	62
18	380	61.7	61.9
18-T	407	62.4	62.3
18-T	405	62	62.8
19	415	62.2	62.6
19	415	62.5	62.7
19-T	404	61.9	62.2
19-T	407	62.2	62
20	335	59	59.2
20-T	388	60.8	60.5

Specimens marked A and B represent the two specimens used for each determination. There were two determinations for each grain size, except for number 20.
The hardness readings for each sample were consistent, all checking within one Rockwell division.

recommendation with reference to determining the area of the flattened surface was followed; that is, the maximum width times the original length of the cylinder gives the area of the flattened surface. In Table VI of the lecture is given the average mutual indentation hardness and in Table XVI are given the individual mutual indentation hardness numbers and the Rockwell hardness numbers of the specimens after being subjected to the temperature of 1200 degrees Fahr. for mutual indentation testing. The Rockwell hardness reading is the average of three. The original hardness of the specimens will be found in Table IV of the lecture. Fig. 56 is the picture of a typical specimen after testing, taken at a magnification of 3.6.

Grain Size

The TaC used in the experiment analyzed 6.22 per cent carbon. Since the residue showed 6.14 per cent carbon and 89.78 per cent

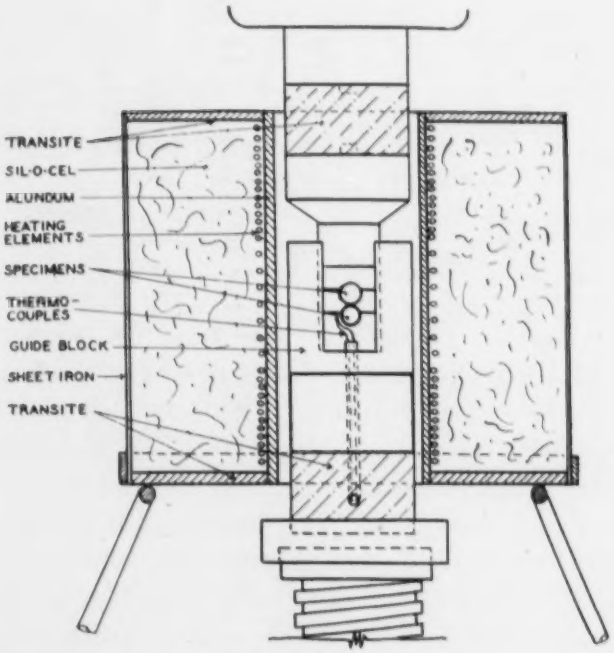

Fig. 55—Detail of Testing Assembly.

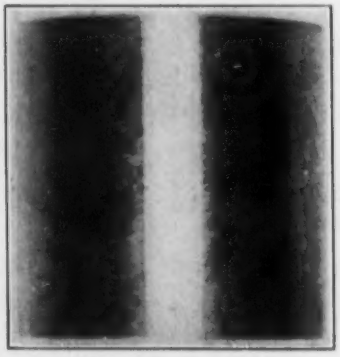

Fig. 56—Photograph of a Typical Specimen after Testing. × 3.6.

tantalum, it seems that dissolving highly alloyed steels in acid to obtain carbide residues for examination and analysis has merit and may not deserve the criticism this method has evoked.

Grain Size and Furnace Atmospheres

For the CO_2 and CO atmospheres the source of gas was bottled commercial carbon dioxide. Carbon monoxide was generated

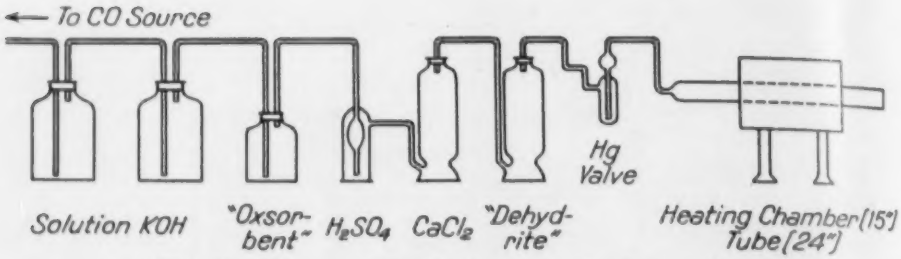

Fig. 57—Diagram of Purification Train for Preparation of CO.

from this by passing through a tube 4 inches in diameter by 18 inches long filled with fine charcoal and heated to 2000 degrees Fahr. in a molybdenum-wound furnace. The resulting CO was purified by carrying through the train of chemicals as shown in Fig. 57. The two bottles of KOH absorbed all traces of CO_2 while the "oxsorbent" absorbed any trace of oxygen. The sulphuric acid, calcium chloride and "dehydrite," of course, were used as driers to remove any traces of moisture. The mercury valve was inserted to prevent back pressure from introducing air into the tube furnace. The purity of the CO gas was determined by actual analysis of samples taken from the tube furnace.

Carbon dioxide gas was purified in the same manner, but, of course, the two bottles of KOH were removed from the "train." The purity of this gas was also determined by chemical analysis.

The third part of the experiment consisted of heating specimens in air. Air was supplied by a small pump and sent directly through the tube with no attempts at drying or purification.

Temperature measurements were made by means of platinum thermocouples. These were calibrated before every heat and the three couples checked within two degrees at all times.

Symposium on the Plastic Working of Metals

Presented before the Eighteenth Annual Convention of the Society held in Cleveland, October 19 to 23, 1936

LAWS AND FUNDAMENTALS OF PLASTIC DEFORMATION

By A. V. de Forest

Abstract

This paper reviews the difficulties involved in setting up either laws or fundamental knowledge in the field of plastic deformation. In the simplest possible case, that of single crystals, the outward mechanism and the results of deformation are known. As yet the reasons for the location of slip planes relative to each other is not understood or the factors which limit the extent of slip on a plane.

The creep behavior of metals at elevated temperatures has long been studied, but the metallurgical reasons for the different behavior of similar material is not clear.

The effect of large deformations is briefly discussed.

IT IS somewhat more than a misnomer to speak of laws of plastic deformation, when present day knowledge covers merely a collection of special cases. Results of plastic behavior in many fields and applications are covered by the great variety of papers presented in this symposium, but it would require the authority of the Medes and Persians to produce a law which "changeth not" and is common to all these fields.

Whether our present knowledge can be termed fundamental will be left to the judgment of the individual student of plasticity. I only predict that our views as of 1936 will be as dated as our present automobiles when another decade has gone by. It is however important to sum up our position and the problems immediately ahead.

The basis of our knowledge is to be derived from the study of the simplest case, where chemical purity and physical homogeneity are at their maximum. The ultimate degree of orderliness of structure occurs in pure single crystals, and we may well pause and consider the outstanding characteristics of such materials in the condition

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A paper presented as part of the Symposium on the Plastic Working of Metals. The author, A. V. de Forest, is professor of mechanical engineering, Massachusetts Institute of Technology, Cambridge, Mass.

in which present refinements of laboratory technique have made them available.

The most striking characteristics are that many of the physical "constants" depend on orientation, the modulus of elasticity of iron for instance varying from about 19,000,000 to 40,000,000, the ductility and proportional limit of aluminum vary by a ratio of about 2 to 1, and the mechanical properties of zinc by an even larger ratio. Some optical, chemical, magnetic and certain electrical properties are all dependent on direction, so that any measurements must be related to orientation before they have any valid significance. At the same time many characteristics, particularly those of strength, are extremely sensitive to small changes in chemical purity, and the importance of this factor increases rapidly as the material approaches absolute purity and homogeneity.

As far as mechanical properties are concerned, it is self evident that any measurements on wood must be related to the direction of the grain. In metals the orientation of structure is less evident, and outside of the field of physics has not been fully appreciated. Since the development of X-rays, a voluminous study has been made of these factors, as summed up in the excellent paper in this symposium by S. L. Hoyt, and this work is being rapidly extended.

The mechanism of plastic deformation is essentially one of shear along crystallographically determined planes. What factors locate the first slip is unknown. As more deformation is imposed, other, usually neighboring but not adjacent, planes also slip. Here are now two most important points, (both unknown). What force and what mechanism stops the slip on the first plane, and what determines the position of the second plane? It is to be noticed that this second plane is very far removed in terms of the lattice dimensions from the first. Quite evidently the total deformation depends on the amount of slip available on each plane and the distance between them. The strengthening, due to deformation, includes both the slip plane and the orderly material between adjacent slip planes. The explanation of all this will probably be found from single crystal study, but the commonly known effects of this change are most evident and familiar in ordinary polycrystalline metals. However, in both cases plastic deformation means slip on shear planes, increase in strength on the plane, the passing on of deformation to other planes strengthening these until the available material is used up and rupture occurs. The relation of deforming forces to crystal

orientation, temperature, and rate of deformation are major variables, and the condition of the crystal as regards composition and previous deformation history are of the utmost importance in the result.

With this briefest reference to fundamentals, let us turn to the rich field of empirical knowledge available in the papers forming this symposium. Plastic deformation is immediately and intimately related to every problem of strength of metals, for in all our mechanical tests the effect of deformation is to a greater or less degree added to the intrinsic properties we wish to measure. This is particularly true of indentation hardness and here some attempts have been made to separate the hardening due to deformation from the hardness due to the previous structure. This change under test conditions is particularly important in the field of what may be called micro-deformation, creep and damping capacity. Creep phenomena are so important from a commercial as well as theoretical point of view that a large group of specialists are constantly adding to our understanding of this phenomenon. Up to the minute information is well summed up in the two papers on the subject. The appalling difficulties of creep testing, involving as it does the utmost refinement of deformation measurement, combined with accurate loading and extremely sensitive temperature control for months and years of time, make any test procedure which can illuminate the same field in a shorter time a matter of great importance. The paper dealing with damping capacity is therefore especially timely. It is reasonable to suppose that a relation exists between this manifestation of micro-deformation and both creep, a term applying to slow changes at constant load, and fatigue, the corresponding factor under repeated load. Damping capacity is a newcomer in the field of testing materials, but it shows possibilities as a test for commercially important factors of strength as well as having great interest from the theoretical point of view.

Leaving the field of very small deformations, we come to an important application of cold work just above the proportional limit. The system of auto-frettage makes use of the relatively large increase of proportional limit brought about by a small permanent deformation of the same type as is to be resisted in service. The resulting system of residual stresses is also a strengthening factor. We may expect this type of treatment to be used elsewhere. For instance, the corresponding method of cold compressing steel tubing

to resist hydrostatic pressure on the outside is now well developed. In spite of the length of time during which metallurgists have known the strengthening effect of over-strain followed by low temperature reheating, there is still no completely satisfying picture of the changes involved. However, at least two major effects are acting in polycrystalline metals. One, the readjustment of internal stresses among the different grains which by virtue of their orientation have different resistances in different directions; and two; the healing, or reconstruction process which takes place in freshly deformed metal crystals under the influence of time and temperature, in other words, elastic recovery. This process is very closely related to precipitation hardening, and may conceivably be the same thing in another form. That is, the keying material instead of being a foreign substance precipitated out of the matrix as in orthodox precipitation hardening, may be a small "fragment" of crystal of different orientation but the same composition acting as a slip interference factor between sheared planes.

The third and most important field is in the region of large deformations such as produced during a great variety of manufacturing operations. Here a division may be made between the desirable properties of metals for hot and cold forming, rolling, forging and extruding, and the effects of such operations on the resulting product. Included in the symposium are papers dealing with many of the endless ramifications of both sides of the problem. Perhaps the most glittering generality, common to all phases of large distortion, is the growing body of knowledge concerning the fibering or reorienting effects of deformation. Cold work and hot work both lead to a structure of preferred orientation of the individual grains, so that as in single crystals there are frequently pronounced directional effects. This has long been recognized, particularly in forgings, where longitudinal, transverse and radial tests have been made since the days of wrought iron, but the quantitative study of these effects, their possibilities and control are now in process of development. Quite recently the great effect of orientation on magnetic properties has been turned to good account. Unfortunately the electrical conductivity of cubic crystal systems does not appear to follow the same course. In the study of these orientation effects, X-ray analysis is of the utmost importance and we may confidently look forward to more information in this field.

The comparatively recent developments in spectral analysis,

coupled with improved methods of refining metals, lead both to better control of important impurities and a higher degree of purity where mechanical properties justify the expense. This process is already under way. Although accomplished without novel methods of analysis, a paper at this meeting, but not in the symposium, shows startling differences in the strength-ductility ratio of such a well known alloy as low carbon 18-8, due to careful control of composition and cold work.

One of the outstanding properties of steel is its characteristic ability to endure great amounts of deformation both hot and cold during manufacture, and then harden by heat treatment either in its final shape, or so close to dimensions as to admit of grinding to finished form. For instance in the manufacture of balls for bearings, cold headed, hardened and ground. With the increasing use of precipitation hardening alloys, many shapes can be formed of ductile metal and then hardened to relatively great strength but lowered ductility.

In the ordinary tensile test, two manifestations of ductility are measured, the percentage elongation and the reduction of area. To a great degree these factors are independent, the relation between them depending on the rate of hardening on one set of slip planes, and thereby passing on the deformation to other planes before the load on the reduced section has reached the breaking point. We have seen before that this hardening is sensitive to both rate of deformation and temperature.

In both tension and torsion test a metal may show great local ductility but only a small elongation over a considerable gage length. The condition is quite different where the deformation is naturally limited by the test conditions, as in the wrapping test for wire. It is possible to have wire with only 6 per cent elongation in 10 inches wrap around its own diameter, in which process the local deformation is of the order of 50 per cent elongation. Quite evidently many manifestations of ductility are possible, and test methods frequently are inadequate to distinguish between the variations which are brought out in service.

A feature of the stress-strain frequently ignored, is the use of the relation between the maximum load and the area of necked down metal which actually supports the load. This figure, frequently called the "true strength" is important as indicating the self hardening under deformation at some unknown temperature, for at normal

rates of loading, the temperature may be significantly greater than that of the surroundings. In impact testing this feature is even more pronounced. The true stress-strain diagram is seldom plotted but appears to have great possibilities especially in the quantitative study of the properties of materials to be subsequently cold formed.

Many of the various manifestations of plastic ductility are described in the group of papers forming this symposium. Much more is in the experience of practical men in daily contact with manufacturing processes, much remains to be studied under scientifically controlled conditions, and lastly, an adequate systematic correlation must be made before the "laws and fundamentals" can be reduced to printed form.

METALLIC SINGLE CRYSTALS AND PLASTIC DEFORMATION

By S. L. Hoyt

Abstract

The metallic single crystals are studied because they enable the investigator to get close to important points of metallic behavior. Until recently the only literature on this subject was scattered through many publications so that the general metallurgist who wished to be posted on the results would set himself an exceedingly difficult task. Fortunately this field has recently been enriched by two excellent treatises, one, "Distortion of Metal Crystals," by Miss C. F. Elam, and the other, "Kristallplastizität," by E. Schmid and W. Boas, whose names will appear repeatedly in any discussion of single crystals. Unfortunately for the present writer these books came to hand after the present paper was practically completed, but not too late to commend them to interested readers for detailed accounts. The present paper attempts to furnish a resume of single crystal literature which deals with selected sections of the general problem of the plastic deformation of metals.

For the purposes of a discussion of plastic deformation we have a very peculiar interest in single crystals for some of the most significant features of our problem are thrown into sharp relief by their use. This is now well recognized and is responsible for the popularity of single crystals in scientific experimentation. By proceeding from the simple to the complex, the factors of grain boundary conditions, varied or random orientation, and constraint of movement of the polycrystalline state can be more clearly differentiated from the factors of crystalline slip, twinning, and strengthening or work hardening.

Production of Metallic Single Crystals

HISTORICALLY the first work on single crystals is rather obscure and was limited to isolated, though significant, observations on native metal crystals and on large grains which were removed from severely coarsened metal parts. Of more recent date

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numerous methods have become available for the production of metallic single crystals in more appropriate form. These methods have the common formula of first producing a nucleus, and then causing crystallization or growth to occur slowly as an extension of the original system. This nucleus grows on more or less completely disorganized material, which may be either a gaseous or a liquid phase or a plastically deformed metal. In this process it is necessary to maintain conditions which do not spontaneously produce nuclei.

W. E. Ruder of the General Electric Company (1913-1914) made use of the principle of straining strips of silicon steel and then annealing them to permit large grains to grow. This was itself an outgrowth of the work of Professor Sauveur which had just previously been published. Later Ruder passed the strained strips slowly through a furnace whose temperature was set above the recrystallization temperature. The preliminary strain set up a definite, though small, tendency to recrystallize in all the grains of the strip. The front end of the strip produced a new crystal on entering the hot zone and it continued its growth along the length of the strip until the whole was converted into a single crystal. Carpenter and Elam used the first method to make single crystal bars and strips of aluminum and published the first complete account of the principles of the process (1). With bars of polycrystalline aluminum of about the correct grain size and properly strained, they found that very slow heating in a muffle furnace to well above the recrystallization temperature produced large single crystals at least part of the time. According to their explanation, all of the grains were strained sufficiently to induce recrystallization on annealing but, by more or less fortuitous circumstances, one grain was strained more than the rest. This grain recrystallized first and then grew throughout the length and breadth of the bar. This method has since been popular for aluminum, magnesium and iron.

A more direct method of "fishing" a crystal out of a melt was used by Czochralski (2). Upon allowing the lower end of a capillary tube to dip into the surface of the molten metal, a small amount is sucked up into the tube. As the hoisting mechanism raises the tube slowly, the metal in the capillary solidifies to a tiny crystal and continues to grow as more and more molten metal is brought up out of the melt. The metal solidifies at the freezing point (no undercooling) at which temperature nucleation does not occur spontaneously. This procedure gives a wire-shaped crystal about 1mm

in diameter and has been popular for the low melting metals like lead, tin, zinc and bismuth.

Another popular method is the one devised by Bridgman (3). The metal is first melted in a tube whose lower end is drawn out to a sharp point. As the mold is lowered slowly out of the hot zone of the furnace, a crystal forms at the tip end and grows along the tube through the entire mass of metal. This method has been used with metals with melting points as high as that of copper by using graphite as a crucible. Both of the latter methods utilize a rate of travel which is less than the velocity of crystallization of the metal, and in fact Czochralski was initially interested in measuring that velocity (4).

Edwards and Pfeil made crystals of iron by the method of Carpenter and Elam by decarburizing mild steel in moist hydrogen, straining slightly, and then annealing 72 hours in dry hydrogen at 880 degrees Cent. (1615 degrees Fahr.) (5). Curiously the simple expedient of allowing the Ar_3 point of iron to pass slowly along a sample has been only little used though the writer made use of it in his work on single crystals (6) as did McKeehan (7).

This is by no means a catalog of methods for making single crystals. Its purpose is to bring out the principles involved and to give the more popular methods by which crystals can be made. The details of these methods and additional methods can be secured from the literature (8).

The experimenter in this field soon discovers a striking characteristic of the single crystal state for which his experience with the common metals does not prepare him,—that being the extreme ease with which the samples may be bent or deformed. A ¾-inch copper crystal 1 foot long can be bent into a horseshoe shape in the hands, while a common zinc crystal will bend under its own weight. This behavior reflects the low yield point and, in mechanical testing, accounts for the extraordinary precautions that are required to preserve the crystal unharmed by the necessary manipulations.

Mechanical Tests of Metallic Single Crystals

A complete account of the properties and behavior of a single crystal requires the orientation of the lattice in the virgin crystal and after various deformations. The best and most reliable means of accomplishing this is by the use of the methods of X-ray crystal analysis, though microscopical or optical methods have also been

frequently used. The latter will be briefly mentioned later, though those who are interested in these methods may refer to the literature on this subject (9). We shall not be able here to discuss methods in detail.

The tensile test deals quantitatively with elastic behavior, plastic yielding, deformation, and rupture in about the same way for single crystals as it does for the common polycrystalline metals. On the other hand, the results differ in an important particular. The test of polycrystalline metals summates the behavior of the individual grains and of the grain boundary effects to such an extent that the normal fine-grained metal reacts much like an isotropic material. In other words the behavior is governed by the applied stress system. The single crystal shows marked anisotropy in both its strength properties and in the way it deforms under the imposed stress system. The bar or wire contracts much more in one direction than in others, and instead of breaking with the usual conical fracture, it either necks down to a sharp chisel edge or fractures along a cleavage face. The effect of the applied stress system is secondary and the behavior of the crystal is governed by the geometry of its lattice. Furthermore, the single crystal is considerably weaker and in some instances is much more ductile than a fine-grained sample of the same metal.

The large number of metals that have been studied in the single crystal state makes it possible and desirable to cover the field with representatives of the three major crystal structures of the common ductile metals; the hexagonal close-packed (HCP), the face-centered cubic (FCC), and the body-centered cubic (BCC). (It is not possible to follow single crystal behavior without an understanding of the simpler types of crystal structure or atomic arrangement and it is assumed that the reader is adequately informed on this phase of our subject.) In carrying this out the author has been guided by the desire to keep the discussion simple so that the principles of plastic deformation may be understood by those who have had no actual experience with single crystals, and yet with a minimum of confusion between experimental findings and hypotheses. This led to the choice of the tensile test for extended treatment because, aside from being familiar to all, it brings out many of the most significant features of plastic deformation. Of the other mechanical tests, the torsion fatigue test in the hands of Gough and co-workers has given illuminating results while compression, plane parallelopipedal com-

pression tests, bend tests, etc., have brought out about the same characteristics of single crystal behavior as the tensile test has.

Hexagonal close-packed metals. Zinc may be used as an example of the HCP metals with magnesium and cadmium as other representatives. Work on zinc, and indeed modern studies of metallic single crystals, started with the classical work of Mark, Polanyi and Schmid. It consisted of tensile tests of single crystal wires of various orientations with observations on the mechanism of the deformation process and the related hardening or strengthening effects. This work showed how strictly the movements in the crystal follow the geometry of the lattice with mathematical precision with the densest packed plane as the one which supports slip and the densest packed direction as the direction of slip (10).

The HCP arrangement has only one densest packed plane, the

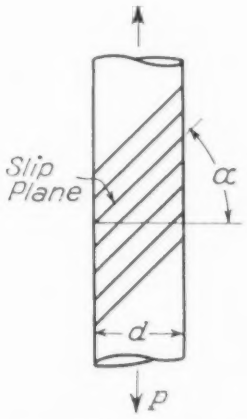

Fig. 1—Representation of Slip Planes in a Single Crystal—Side View.

basal plane (0001). This suggests the homely analogy of a deck of cards to represent the zinc crystal with all slip planes parallel to each other. When they lie at an angle to the wire axis as shown in Fig. 1, a tensile pull stretches the wire by slip along these planes, and during the extension they tilt towards the long axis,—see Fig. 2. This causes the wire to stretch out into a thin ribbon or band with an elliptical cross section. As soon as the angle of the basal plane reaches 70 to 80 degrees the crystal generally fails by cleavage along the basal plane or on one of the three prismatic planes. If the basal planes are originally either parallel to or normal to the wire axis, no slip occurs and the wire fails by a brittle cleavage on a prismatic

or basal plane. Actually the basal plane must start out a few degrees away from the parallel or normal position for the crystal to deform plastically, but with the latter orientation the greatest possible ductility is obtained with elongations of several hundred per cent.

With the initial orientation known, it can be readily shown that there is a fixed relation between the elongation of the crystal and the increase in tilt of the basal plane. Calling this angle alpha of Fig. 1, the elongation is related to the tilt of the basal plane by the cosine formula

$$\frac{l_1}{l_o} = \frac{\cos \alpha_o}{\cos \alpha_1}*$$

in which "l" is the gage length and alpha is the angle of the slip plane, "o" signifies the initial values and "1" the values after the crystal has stretched.

The slip ellipses which appear on the zinc crystal are distinct and continuous around the crystal. Their angle of tilt can be followed approximately with the telescope of a cathetometer which is used to measure the length of the crystal, l. The more precise methods of X-ray crystal anaylsis make it possible to follow this angle from the initial orientation on.

We have mentioned that slip is restricted to the direction of closest packing in the slip plane. Slip along the basal plane of zinc (0001) is restricted to one of the three digonal axes, such as $[1\,1\,\bar{2}\,0]$ and with these three axes assuming any position in the slip plane, it will be the one which most closely coincides with the direction of maximum resolved shear or the major axis of the elliptical section. By resolving the shear stress along the next adjacent digonal axis, it becomes clear that a moment is set up which causes the basal plane to rotate so that the slip direction comes more and more into coincidence with the direction of maximum shear. This produces a rotation about the hexagonal axis and on this account the cross section increases in width and becomes eccentric. In the vicinity of the grips this behavior begins to disappear and it finally vanishes at the grips where the crystal is held rigidly. The relation between

*Complete formulas for the problems of deformation are given in the books of Elam, and Schmid and Boas. The working formula for elongation is either

$$\frac{l_1}{l_o} = \frac{\sin \lambda_o}{\sin \lambda_1}, \text{ or } \frac{l_1}{l_o} = \frac{\sin X_o}{\sin X_1}$$

where λ is the angle between the wire axis and the direction of slip, and X is the angle between the wire axis and the slip plane.

the direction of maximum resolved shear in the slip plane and the direction of slip in that plane is shown in Fig. 2.

We have briefly covered the crystallographic mechanism of slip during the tensile extension of ductile zinc crystals, and have now to consider the important stress-strain relationships of the tensile test. By referring to Fig 3 it is clear that as soon as the load exceeds a certain value, the crystal commences to deform rapidly. This point on the stress-strain diagram is the yield point and by reference to Fig. 4 it is clear that differently oriented crystals yield at different values of the stress. It has been noted by Schmid that the resolved shear stress acting along the slip plane at the yield point is constant

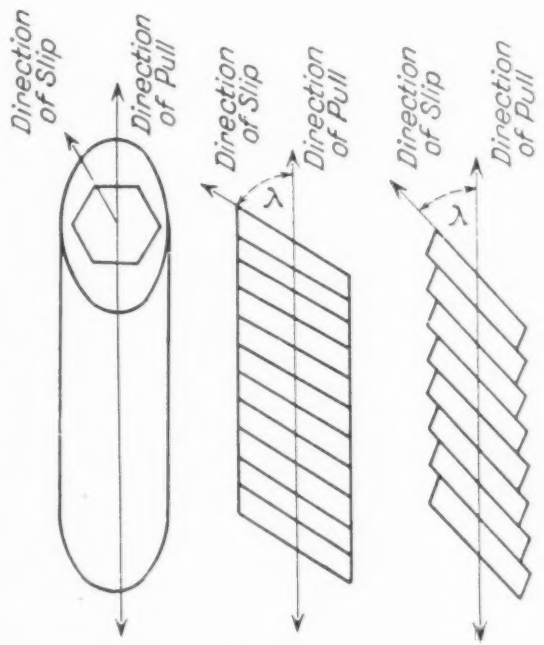

Fig. 2—Representation of Direction of Slip in Basal Plane of Zinc.—Elam. (Ref. 10, from Elam, Ref. 9.)

and independent of the initial orientation, while Gough has observed the same thing for crystals tested in torsion. This is known as the "critical resolved shear stress" principle and according to it the crystal begins to yield to the applied stress as soon as the resolved component reaches the critical value. It must not be assumed, however, that no plastic deformation occurs at lower stresses for more precise determinations have shown that some deformation does occur at stresses well below those required to produce marked yielding.

Presumably the critical shear stress law would also apply to the early stages of deformation as well.

As soon as slip begins, the resistance to further slip increases. It is said of this behavior that the crystal "strengthens" or "work hardens." This is shown in Fig. 3 by the stress-strain diagram at

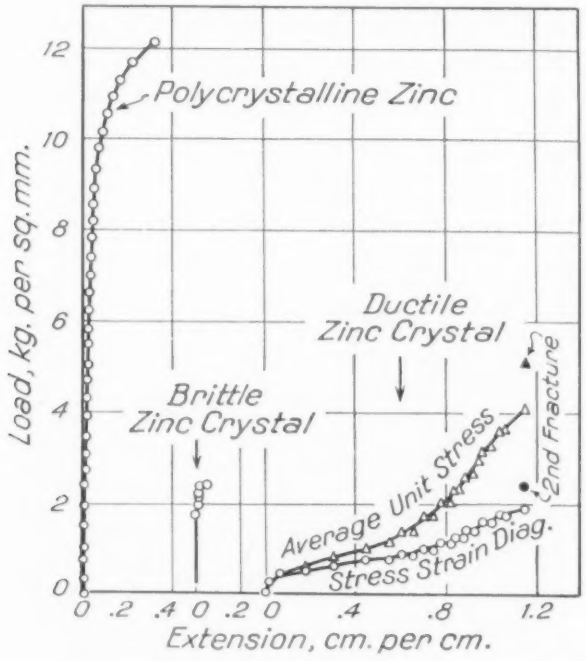

Fig. 3—Stress-Strain Diagrams of Zinc. (1) Polycrystalline Zinc. (2) Brittle Zinc Crystal. (3) Ductile Zinc Crystal.

(3), while the stress-strain diagrams of crystals with other initial orientations are shown in Fig 4a. From the appearance of the stress-strain diagrams one would not realize that the metal itself is reacting in the same way in all of the crystals. This comes from plotting the stress as the load divided by the initial cross section and the elongation as per cent elongation for a given gage length. We secure a much better understanding of both deformation and strengthening by plotting the shear stress on the basal plane against the shear along the slip plane, as is done in Fig. 5 for the crystals of Fig. 4. The shear is taken as the relative displacement of two slip planes which are unit distance apart. This stress-strain relationship is found to be independent of the initial inclination of the slip plane. Instead of giving a sheath of curves, such values combine to give just one curve to within about 10-15 per cent. This type of curve is the

"shear stress-shear" curve and best characterizes deformation and strengthening of single crystals. It has no counterpart in tests of polycrystalline metals, but is obviously the proper curve to use for comparisons of different metals or to secure the relative behavior of the same metal when tested under different conditions. We should

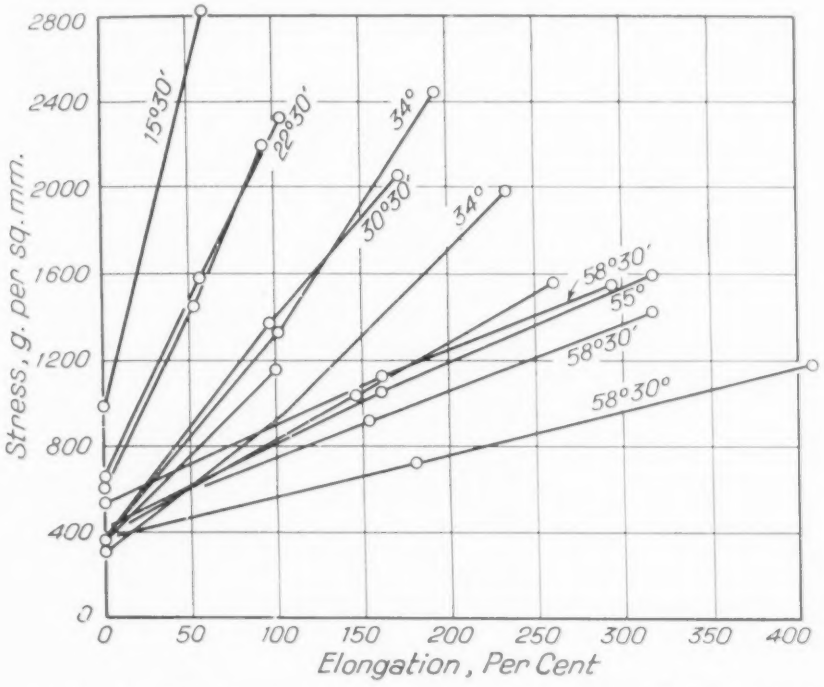

Fig. 4a—Stress-Strain Diagrams of Zinc. Crystals with Various Initial Orientations of the Slip Plane. The Angle X_o is Given. (See E. Schmid, Ref. 28).

note in particular that the usual determination of the elongation as

$$\frac{l_1 - l_o}{l_o}$$

may be misleading in single crystal work because a given extension may correspond to quite different amounts of shear, depending on the orientation of the slip planes.

It has been noted that the shear stress which is required to initiate slip on the basal plane is constant and independent of the inclination of the slip plane. This is surprising for it can be true only if the normal component of the load acting to separate the slip planes by cleavage, is without effect. A specific test of this point was made by Polanyi and Schmid by determining the shear stress at the yield point when external pressures up to 40 atmospheres were

applied, but the results were negative (11). Rosebaud and Schmid also determined the yield point for different angles of the basal plane and found that the load was symmetrical about the 45-degree angle even through the normal component was quite different with larger and smaller angles than 45 degrees. Actually the shear stress

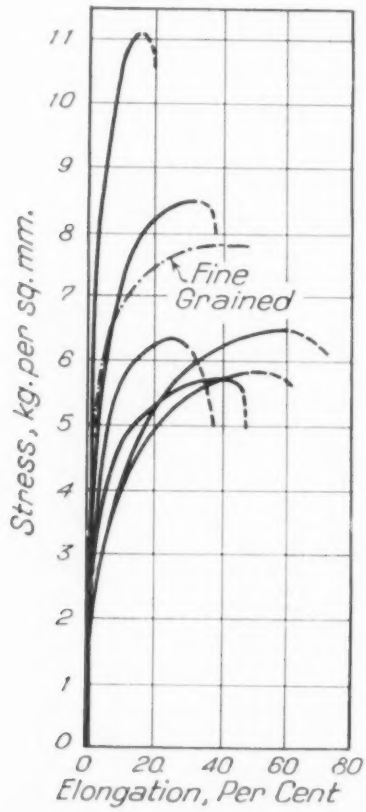

Fig. 4b—Stress-Strain Diagrams of Aluminum. Crystals with Various Initial Orientations of the Slip Plane. (See R. Karnop and G. Sachs, Ref. 27).

remained constant in spite of a change in the normal component of 50:1 (12).

In this discussion it has been intimated that the angle alpha increases continuously throughout the deformation and that this occurs simultaneously over the whole length of the crystal (uniform elongation) and that the cosine law holds for all values of the elongation. That is what the present author has observed (68). According to the original description of Mark, Polanyi and Schmid, the behavior was quite different in that deformation started locally by the slip

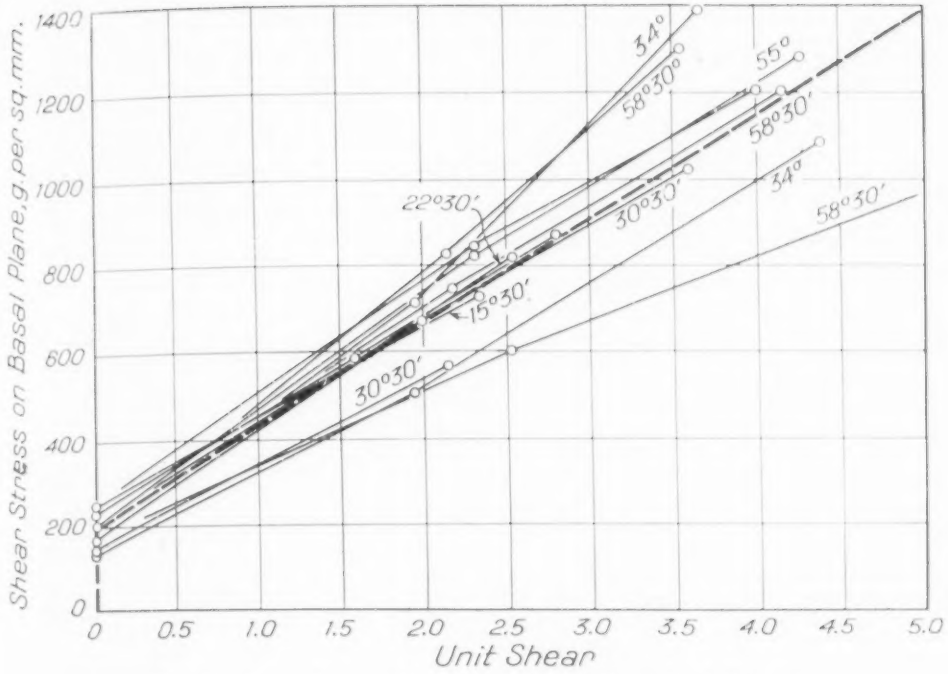

Fig. 5a—Shear Stress—Shear Diagrams of Crystals of Fig. 4a.

planes at that place going suddenly and completely from the initial to the final position. Further deformation took place by spreading from this place along the length of the crystal. Schmid has shown splendid examples of the spreading type of elongation in cadmium crystals, while this same type of behavior in zinc has been described by Miller (13). At present it is not possible to account for this difference in behavior.

Before leaving zinc the interesting phenomenon of "after elongation," first described by Mark, Polanyi and Schmid (10), will be briefly considered. After the crystal wire has elongated to a thin elliptical ribbon during the primary elongation, the crystal may deform anew at some local place by marked constriction of the ribbon in the direction of the major axis. This results in the formation of a fine thread of cylindrical shape whose diameter is the same as the minor axis of the elliptical ribbon. The elongation represented by this thread is several hundred per cent and the strengthening may be as much as fifty fold. There has been some difference of opinion as to the mechanism of this effect and the atom planes involved in the movement. At first it was held to be due to slipping on a prismatic plane (10) but Matthewson and Phillips pointed out that

this movement must be preceded by twinning about a first order pyramidal plane (1012) which would place basal planes again in position for slip (14). The new position of the basal plane comes within 4 degrees of the position of the prismatic plane, or close enough to make it difficult to distinguish between the two. While the reason for such a twinning movement is not at all clear, striking microscopic and X-ray evidence of the correctness of the idea was given by Schmid (15) though Wilson and the writer were unable to

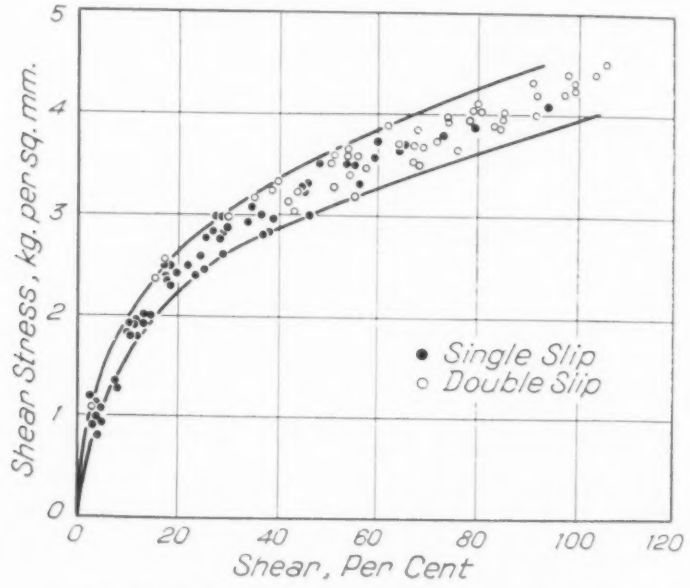

Fig. 5b—Shear Stress—Shear Diagrams of Crystals of Fig. 4b. Dot = Single Slip. Circle = Double Slip.

detect this type of movement except in larger crystals (16). When Schmid corrected earlier data on the strengthening of zinc crystals it appeared that the resistance to slip on the basal plane increased materially during the twinning movement. Afterwards there was a slow decrease in strength during the after-elongation stretch.

So far in our discussion of the strengthening effect of deformation we have been dealing with the resistance to deformation; now we should note that deformation may also increase the resistance to rupture or cleavage. Schmid studied this point by pulling crystals at room temperature up to the breaking point and then again determining the breaking strength at the temperature of liquid air. With fracture occurring on the prismatic plane, the strength was found to increase with the previous deformation at room temperature from

2.6 to 9.8 kilograms per square millimeter. The room temperature deformation varied from 1.5 to 6 fold depending on the initial orientation. It will be noted that the prismatic plane crosses the planes which supported the slip; when fracture came on the basal plane no such increase in cleavage strength was observed (17).

An excellent exposition of the deformation of hexagonal crystals is that of Boas and Schmid for cadmium (18). This gives a general confirmation of the previous findings with zinc, except that cadmium may become weaker with the first slip. For other work on hexagonal metals, reference will be made to the literature (19).

On passing to the FCC metals we may be guided by the general principles which are brought out by the work on zinc. With one slip plane, zinc serves as an excellent introduction to single crystal deformation and the modifications introduced by the multiple possibilities of slip of the FCC lattice should be readily understandable.

Face-centered cubic metals. Historically it was early proposed by Mügge that one mechanism of deformation was that of "gliding" or translation of portions of the crystal over each other, the other mechanism being twinning. His studies were made on crystals of native copper and gold, both FCC. In what was probably the first metallurgical work on metallic single crystals, Humfrey studied the deformation of lead and observed by the aid of etching pits that slip occurred parallel to the octahedral or (1 1 1) planes (20). He also observed that the crystal did not remain perfectly uniform in orientation, and even that new orientations were produced, of which some were in twinned positions. Additional early work on single crystal wire of lead, tin and mercury, was that of Andrade who observed the change to elliptical cross section when the wire was slowly stretched out (21). This early work was lacking in an analysis of the atomic mechanism of the deformation, for which we must turn to more recent work.

In the paper of Carpenter and Elam already cited, an introduction was given to the behavior of aluminum crystals in the tensile test (1). It was shown that the single crystals varied from 6600 to 8800 pounds per square inch in tensile strength and from 55 to 87 per cent in elongation in 3 inches, while the fine-grained metal ran 10,600 pounds per square inch and 44 per cent respectively. This relatively small variation found with single crystals of aluminum is characteristic of the more symmetrical structure. Later Miss Elam took up a thorough analysis of the deformation and rupture of alu-

minum single crystals with Taylor and the following description is essentially the one they give (22). The methods of analysis used for zinc are not satisfactory for aluminum and Taylor and Elam made use of the principle of the unextended cone in their work.

The planes of densest atomic packing of FCC metals are the octahedral or (1 1 1) planes of which there are four equivalent sets which can function as slip planes. Each one of these sets of planes has three directions along which slip can take place so that this

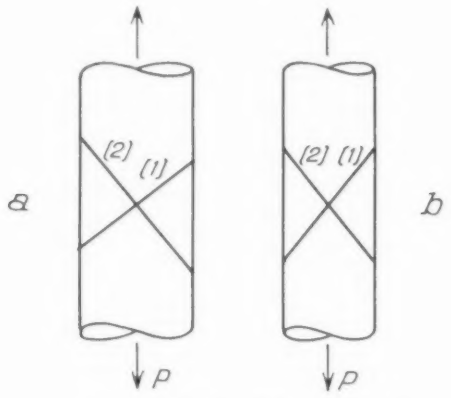

Fig. 6a—Active (1) and Latent (2) Slip Planes in FCC Crystal—Initial Positions.
Fig. 6b—Same as Fig. 6a at Start of Double Slip.

lattice has, in all, twelve directions of slip. When a tensile pull is applied to a single crystal bar the shear stress along one set of (1 1 1) planes will be, in general, greater than along any other set. This general case will be considered first. Slip starts along this plane as soon as the resolved shear stress exceeds the resistance to slip. Following the general rule, the direction of slip is the adjacent [1 1 0] direction. Just as was the case of zinc, the slip causes a rotation of the lattice in the sense of lining up slip plane and slip direction with the axis of the crystal or with the direction of pull. This time, however, the geometry of the FCC lattice is such that this movement brings another set of octahedral planes into an approximately symmetrical relationship with respect to the axis. The result of this is that two sets of slip planes now become active.

The situation at the commencement of slip is shown at (a) in Fig. 6. The octahedral plane (1) has a higher shearing stress than the one at (2) and hence it is selected as the slip plane. During the first or primary slip, the lattice rotates to assume the position shown at

(b), looking at the crystal from the side. At the same time the crystal becomes thinner in the dimension shown. The octahedral planes represented by (2) have also rotated through the same angle so that the area of the active slip plane remains constant while that of the inactive slip plane becomes smaller due to the lattice movements during this stage. At the stage represented by (b) the areas are approximately the same though there is a small difference of a few degrees in the inclination of these two planes. It is evident from the geometry of the crystal at this stage that slip now occurs on both planes simultaneously, though the difference in inclination of the two planes gives a lower stress on the old plane than on the new plane. No further rotation of the lattice occurs and the crystals soon neck down locally to a chisel fracture.

Taylor and Elam followed the increase in resistance to shear and set up a mathematical relationship between the rate of strengthening and the rate of extension. This relation shows that aluminum will never fail by single slip because the conditions of double slip enter before the rate of extension exceeds that required for fracture. It is on this account that aluminum single crystals always show the wedge fracture. An additional and quite detailed account of tests of aluminum crystals of various orientations is that of von Goeler and Sachs (23). The true rate of strengthening, determined as the increase in resistance to shear versus the amount of slip along the slip planes, was found to remain essentially constant for the different crystals and independent of the orientation and normal stress on the slip plane; see Figs. 4 and 5. The shear stress at the point of yielding was not determined but was obviously quite low.

Systematic studies of crystals of selected orientations confirm these findings. The yield point of FCC metals is a minimum for those lattice directions which are 45 degrees to slip plane and slip direction. The maximum yield point is in the direction of the body diagonal [1 1 1], with a ratio of 1 : 1.84 (52). Czochralski gives the proportional limits of aluminum crystals as 850, 1100, and 1700 pounds per square inch respectively for [1 0 0], [1 1 0], and [1 1 1] directions (25). In general the tensile strength varies in about the same way with orientation as the tensile stress required for yielding, with the maximum in the [1 1 1] direction. The maximum elongation comes in the [1 1 0] direction, with minima in the [1 0 0] and [1 1 1] directions (24, 25).

An observation made on aluminum that is but little understood is

that while the deformation described by Taylor and Elam holds for temperatures from —185 to 400 degrees Cent. (—120 to 750 degrees Fahr.), at temperatures above 400 degrees Cent. additional slip occurs on (1 0 0) in the [1 1 0] direction (26). This agrees with the finding that the end orientation of aluminum which is rolled at temperatures above 450 degrees Cent. (840 degrees Fahr.) has [1 0 0] or [1 1 1] parallel to the direction of rolling, depending on the initial orientation (27). Work on iron, which will be discussed shortly, shows that slip may not be confined to the planes of densest packing and in aluminum we may have some such a situation at elevated temperatures.

The stress-strain relationships are about the same here as was given for zinc, except for a marked increase in the stresses and an even more pronounced decrease in elongations; see Figs. 4, 5 and 7. The multiplicity of slip planes which produces ductility in the FCC metals in the polycrystalline state acts to limit the ductility of the single crystal by the intervention of double slip. On the other hand, the inherently more brittle HCP arrangement provides a slip mechanism in favorable cases which permits very large elongations before fracture. It is again found that the shear stress along the slip plane is independent of both the orientation and the normal component of the stress. We also see another principle confirmed which is that the active slip plane is less affected by slip than are other lattice planes which cut the slip plane at an angle (28). The differences between the FCC and HCP metals are brought out by the shear stress-shear diagrams of Fig. 7.

Body-centered cubic metals. In the body-centered cubic lattice we find an arrangement which, on the basis of the principles which we have discussed for both HCP and FCC structures, does not appear to lend itself to deformation so readily. This is reflected in the known low ductilty of the BCC metals tungsten, molybdenum, and chromium and, in fact, this relationship between availability of close packed planes and ductility was early proposed by Hull. Iron stands out as an exception to the rule. Before entering the discussion of representatives of the BCC metals it may be of interest to note what predictions we would make for slip plane and direction of slip for that arrangement. If we reason by analogy we would select (1 1 0) and [1 1 1] respectively, for they are the plane and the direction of greatest atomic packing. Oddly enough, this seems not to be the case and of these two predictions it is only that for the slip

direction that is fulfilled. The slip plane is variously reported as (1 1 0), (1 1 2), and (1 2 3), etc., and the situation is not as yet thoroughly understood. Before passing to the details of the slip mechanism we shall first consider the more common properties of iron crystals.

The most extended investigation of the production and general properties and behavior of iron single crystals was that of Edwards

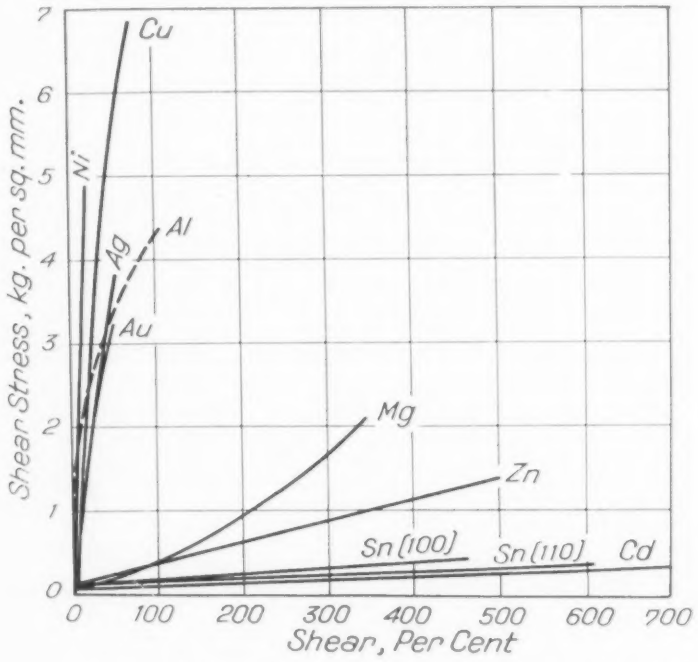

Fig. 7—Shear Stress—Shear Diagrams of Various Metals. (E. Schmid, Ref. 81).

and Pfeil (29), and later continued by Pfeil (30). They showed that single crystal iron begins to deform plastically at much lower stresses than ordinary iron or mild steel does. This particular determination depends on the precision of the measurements of the elongation and we ought to keep this fact in mind on comparing the results of different investigators. Their figures for the two states were about 3800-5500 pounds per square inch and 22,000 pounds per square inch respectively. The tensile strength came about 22,000 pounds per square inch for most of the crystals, or about one half that of common iron, though one crystal of uncommon orientation came at 35,000 pounds per square inch. These figures agree very well with some later work of Ziegler (31). A striking observa-

tion was that the stress-strain diagram of single crystal iron does not show the well known yielding of mild steel, in this respect to act like single crystals of other metals. They also confirmed an earlier observation of Osmond and Fremont (32) that the iron crystals are brittle in one direction and break with a sharp cleavage along (1 0 0). Brittleness of other metals which we have discussed is accounted for in an orthodox manner, but this peculiar brittleness in iron would appear to warrant more study to bring out its true significance.

More recently a series of iron crystals has been tested by Fahrenhorst and Schmid with orientations which practically covered the entire range of possibilities (33). The tensile strengths which they report agree well with those of Edwards and Pfeil, but from the materially higher yield points it may be inferred that their measurements of elongations were somewhat less exact. Without attempting here to correlate their findings with the orientations, they reported yield points from 11,500 to 18,500 pounds per square inch, tensile strengths from 23,000 to 32,500 pounds per square inch, and elongations from 17 to 84 per cent for Armco iron crystals.

Though it is not the present writer's intention to discuss behavior in compression, a brief account of Pfeil's tests will be instructive. (30, first paper) Iron crystals show the highest elastic limit when compressed on the (1 1 0) plane and the lowest on the (1 0 0) plane. The values given are 6500 and 4350 pounds per square inch respectively, though he states that the true elastic limits probably lie 1000 to 2000 pounds per square inch below those figures or at about the stress at which slip bands first appear. Upon being further compressed, the (1 0 0) orientation becomes stronger than the (1 1 0) orientation at about 10 per cent reduction in thickness.

Coming to the deformation of iron crystals, it has always been observed that the slip bands are wavy, whereas in other metals they are straight. This circumstance has caused a lot of discussion and speculation on the atomic mechanism of slip in iron. Perhaps the most reasonable early explanation was that of Rosenhain who pointed out that the crystals of iron, in which the curved bands were observed, had formed in the solid state during an allotropic transformation, and which likely departed from perfection (34). This explanation was logically related to a disorganized movement of some sort, though according to more recent teachings, that would be due to peculiarities of the iron lattice and not necessarily to distortions of that lattice before straining.

O'Neill reported that slip in an iron crystal containing 1.8 per cent silicon occurred on the (1 1 0) plane though later work caused him to change this to (1 1 2) (35). Pfeil carefully examined his compressed crystals for evidence of the slip plane. By judging from whatever slip band evidence was available and from the change in shape of the crystals, he eliminated (1 0 0) and (1 1 0) as possibilities. Further consideration convinced him that the slip plane must be of the type that intersects all three axes. In the cubic system there are two such principal planes, the octahedral (1 1 1) and the icositetrahedral (1 1 2). More careful study eliminated (1 1 1) and he finally concluded that (1 1 2) is the slip plane of iron. (30, first paper). In this work Pfeil made excellent use of pressure figures to secure the orientations of his samples. Taylor and Elam were the first to make a systematic study of iron crystals by X-ray methods and they used the same technique they had developed for aluminum. This work led them to the conclusion that it is only the direction of slip [1 1 1] that is crystallographically determined. Thus it appeared that the slip process in iron was not like that of the deck of cards, but was more like that of a bundle of matches being moved laterally. The plane of slip was uncertain and appeared to vary between (1 1 0) and (1 1 2), depending on the orientation of the crystal or upon the intensity of the resolved stress on the planes (36). It is unfortunate here that a well developed system of slip bands is not available for purposes of checking the approximate position of the slip planes in the lattice.

Another method of approach was that of Wever who found that the end position or preferred orientation of the lattice in deformed iron is the same as that which would be established if slip were to occur in the "orthodox" manner along the (1 1 0) planes in a [1 1 1] direction (37). Gough, on the other hand, and on the basis of torsion fatigue tests of iron crystals, concluded that slip occurs on pairs of planes instead of just one. There were two of these pairs, either (1 1 0) and (1 2 3), or (1 1 2) and 1 2 3), depending on the resolved shear stress acting on the pairs of planes. The direction of slip was taken as the intersection of the pair of planes, in either case the [1 1 1] direction (38).

The latest contribution to the solution of this perplexing problem is that of Fahrenhorst and Schmid in a studied review of the various possibilities of slip in iron crystals. With no possibility of using slip bands for more direct evidence, they followed the changes

in orientation during deformation, the variation of the yield point with initial orientation, the rate of strengthening, and the widening of the crystal during elongation. Upon comparing their findings with the theoretical requirements for various possible slip planes, they concluded that the assumption of slip along (1 2 3) was the most reasonable. Slip along only (1 1 0) could not account for all of their observations so that they show positively that some plane, other than the one of closest packing, is active. Slip along (1 1 2) was consistent with the changes in orientation but would require the crystal to increase in width by about 9 per cent, which was not found. Slip along (1 2 3) was held to agree best with experiment, with no discrepancies. Some evidence of slip on at least two planes was observed by the decrease in width of the crystal, but was apparently held to be a minor effect. The direction of slip was found to be [1 1 1], as before (33).

A few other BCC structures have been studied but not in as great detail as iron. Goucher studied single crystals of tungsten at temperatures ranging from 1000 to 3000 degrees Kelvin, and while his work did not go into detail of other similar investigations, it is clear that essentially the same principles of deformation, strengthening, etc., were observed (39). He reported that slip occurred on two sets of (1 1 2) planes simultaneously and in the [1 1 1] direction. This resulted in the formation of a wedge shaped fracture such as is always found with iron crystals. Taylor studied the BCC beta brass with results which were very similar in many respects to those for iron, particularly with certain orientations. With other orientations he found slip along (1 1 0) in the [1 1 1] direction (40).

The Orientation Effect. Theory requires and experiment shows that the load which produces movement on the slip plane varies with the orientation of the slip plane. This is true because it is only the resolved shear stress which is effective. It is obvious that the change in orientation during deformation must have some effect on the load-elongation relationship, and it may even raise the question of the likelihood of accounting for strengthening on that basis. With single crystals available for study, simple calculations show that this cannot be done for the stress would fall above the yield point if the shear stress remained constant (84). Clearly the deformation process strengthens the crystal and no part of this effect can be accounted for on the basis of the change in orientation during slip.

When we transfer this reasoning to polycrystalline metals, it becomes clear that a psuedo-strengthening is produced by the orientation effect, for there the stress-train relationship is used and not the shear stress-shear relationship. The grains whose slip planes are 45 degrees to the direction of the applied load will be the first to deform and somewhat higher loads will be needed to produce slip in the other grains. It is not my intention to discuss this here, but calculation shows that any "strengthening" which could be accounted for on this basis would be limited to the load required to produce slip on the least favorably oriented grain. This is a relatively small effect and is entirely out of line with the real increase in strength which is observed, though this is frequently overlooked in the literature.

Hardness of single crystals. The methods used to determine the hardness of metals imposes deformations on them, and in view of the variations in behavior between single crystals and polycrystalline metals which we have just reviewed, it will be of interest to note that hardness determinations have shown no such differences. Thus, while O'Neill found the rhombic-dodecahedral face (1 1 0) to give the highest hardness for aluminum, with the octahedral face (1 1 1) next, and the cubic face (1 0 0) next, all three faces came within about 5 per cent of the hardness of the same material in the fine-grained condition (41). Working with iron, Pfeil found about the same situation except that the superior methods of Meyer's analysis, which were appropriately used, showed that there is actually no difference in hardness between the different orientations. The constants found were "a" = 74, and "n" = 2.36 for all lattice directions. He determined the Brinell hardness numbers at the same time and found that they varied from 79.6 to 85, or by more than 5 per cent. When he compared the single crystal iron with the same material in the polycrystalline state with grain sizes varying from "very fine" to coarse, he found very little difference. For all of the polycrystalline samples, "a" was again 74, while "n" was 2.38. The Brinell number varied from 76.4 to 83.4. Pfeil also treats cold working and annealing very fully in his work on iron (42).

Effect of temperature. The effect of temperature on the properties of single crystals is about what would be predicted from the known behavior of polycrystalline metals, with some variations. The figures of Table I give the effect of subzero temperatures on

the shear stress at the yield point of zinc and cadmium as reported by Schmid, Rosbaud and Polanyi (43). Even at a temperature

Table I
Effect of Temperature on the Yield Point

	Shear Stress at the Yield Point gm/sq mm	
Deg. K.	Zinc	Cadmium
300	100	70
83	130	110, 70
20	185	75

of only 1.2 degrees above absolute zero, zinc and cadmium have yield points of the same magnitudes (44), so that the effect of low temperatures on the initial resistance to slip is only of the first order. The slip bands appear at about the same distance apart at all temperatures, except that at elevated temperatures the block movement occurs on a coarser scale.

The increase of stress with deformations above the yield point is low for both zinc and cadmium at room temperature and varies considerably with the speed of straining. The rate of strengthening of aluminum is much more rapid while the speed of deformation has a relatively small effect (45). At room temperature the whole curve is displaced upwards about 16 per cent when the velocity of straining is increased 23,000:1 (85). At low temperatures zinc and cadmium behave more like aluminum at room temperature. For example, cadmium at room temperature shows an increase of the shear stress of 40 per cent at a deformation of 40 per cent, but at 20 degrees Kelvin the shear stress increases by 370 per cent at the same deformation. These points are illustrated by Figs. 8 and 9 which are taken from publications of Schmid. It is evident that at low temperatures the stress depends mainly on the amount of strain. It was found that the same is true at temperatures near the melting point, while the reason for this is said to be that the crystal recrystallizes or "recovers" as fast as it deforms and hence remains in the "annealed" condition (46). Curiously, the experimental data indicate that above 100 degrees Cent. (212 degrees Fahr.) for cadmium, and above 200 degrees Cent. (392 degrees Fahr.) for zinc, the strength becomes essentially unaffected by further increase in temperature. According to this, we may expect the crystal at the melting point to possess a definite amount

of mechanical strength (47). Over the temperature range of —185 to 600 degrees Cent. (—300 to 1110 degrees Fahr.) increasing the temperature lowers the yield point and rate of strengthening of aluminum, though the total shear to rupture remains constant at 120 per cent (26).

Impurities and alloy single crystals. So far we have dealt with essentially pure metals in single crystal form. It has been ob-

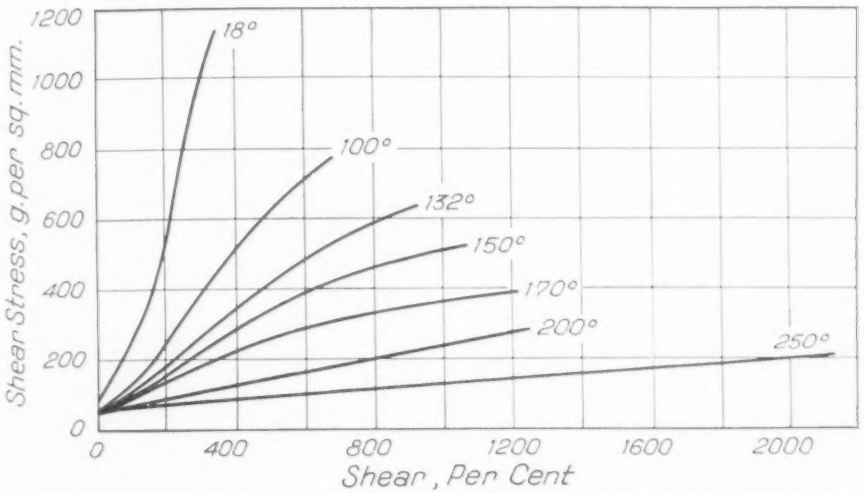

Fig. 8a—Shear Stress—Shear Diagrams of Zinc at Various Temperatures. Degrees Cent. (W. Fahrenhorst and E. Schmid, Ref. 47).

served in such studies that small amounts of impurities have measurable effects on the properties of the crystals. One generality which stands out here is that the purer the crystal, the weaker it is. This is the same effect that is found with polycrystalline metals, particularly with those in which the impurities enter the parent metal in solid solution. This effect is nicely illustrated by the results of two series of Zn-Cd and Zn-Sn alloys given in Table II.

Table II
Shear Strength at Yield Point in Zinc Alloy Crystals

Cadmium-zinc (1)	gm/sq mm	Tin-zinc (2)	gm/sq mm
0.03% Cd	94	0.0% Sn	94
0.13% Cd	214	0.1% Sn	244
0.53% Cd	815	0.5% Sn	313
1.03% Cd	1186	1.0% Sn	322
		1.5% Sn	373
		2.0% Sn	355

(1) P. Rosbaud and E. Schmid. *Zeit. tech. Physik.* 1925, 32, 197.
(2) E. Schmid. *Zeit. Physik.* 1926, 40, 54.

The addition of cadmium to zinc raises the strength over ten fold and the increase is continuous up to the highest amount added. This amount of cadmium is soluble in zinc so that the increase reflects the effect of the cadmium entering the zinc in solid solution. Tin is soluble in zinc to a lesser extent and the rapid initial strengthening effect soon dies out. The tin which is in excess of about 0.5 per cent occurs as a separate phase which, in this case, forms char-

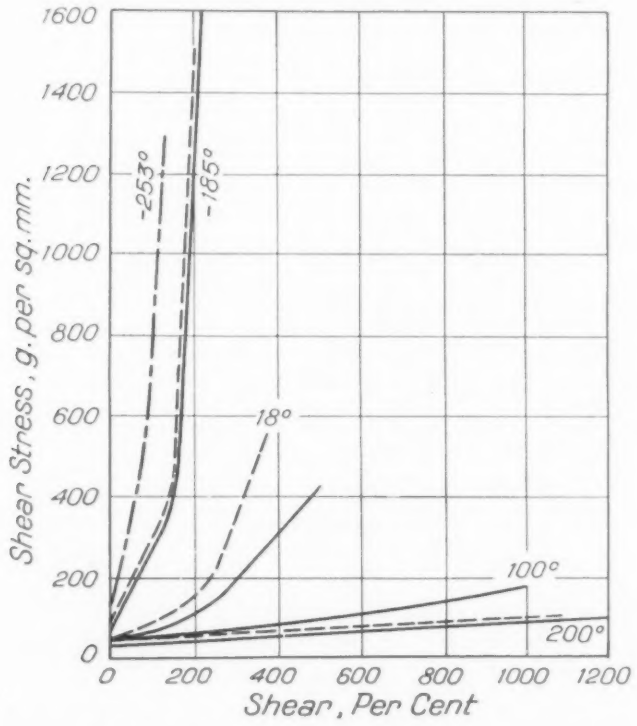

Fig. 8b—Shear Stress—Shear Diagrams of Cadmium at Various Temperatures. (W. Boas and E. Schmid, Ref. 46). Rate of Loading 100 Grams per Minute. With Dash Curves, Rate was 100 Times as Fast.

acteristically as thin layers along the basal planes. The crystal is composed of blocks of the zinc-tin alloy separated by the layers of tin. It appears that these layers are so thin that they produce a "thin layer" effect which enables them to stand appreciably higher shear stresses than massive tin of the same composition could.

Tests by Elam on aluminum crystals with 18.5 per cent zinc in solid solution show that the deformation mechanism remains unaltered by alloying, but that the materially reduced ductility gives rise to a different fracture (48). In this case fracture usually occurs

on the slip plane before the advent of double slip, or else along a (1 1 2) plane. Zinc raises the yield point of aluminum but lowers its rate of strengthening.

The addition of zinc to copper has the same effect of increasing the elongation as it has in common brass. On the other hand, it alters the mechanism of double slip both quantitatively and qualitatively. Instead of slipping simultaneously on old and new planes,

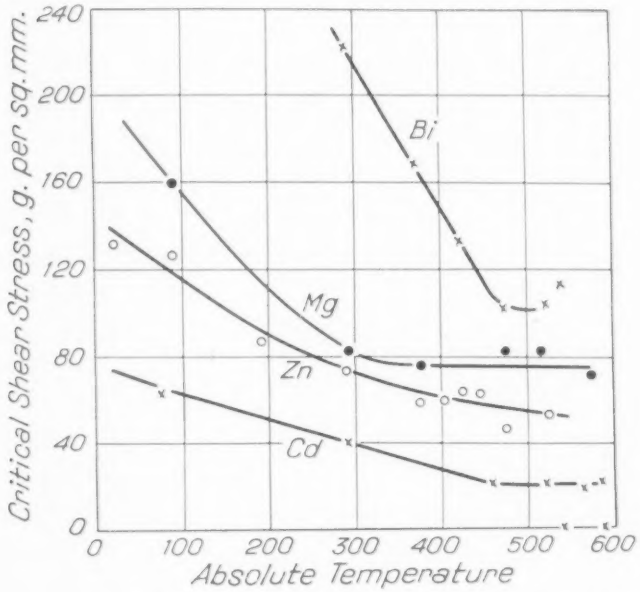

Fig. 9—Effect of Temperature on Critical Shear Stress. (Collected by E. Schmid).

the alloy crystal alternates between the two, while the shear stress on the new plane is much higher than would be expected. Sachs and collaborators found that zinc increases the shear stress at the beginning of elongation from 550 pounds per square inch for pure copper, to 2000 pounds per square inch for the 28 per cent zinc alloy (49). The addition of 5 per cent aluminum to copper makes the copper crystal harder, but again the alloy crystals are more ductile than pure copper. Even though pure copper hardens faster with slip, the greater ductility of the alloy crystals leads ultimately to greater strength (50). Additional work on alloy crystals confirms these results and, in particular, substantiates the point that the alloying metal can materially strengthen the parent lattice without lowering the ductility (51).

Some selected data are reproduced in Table III.

Table III
Tensile Tests of Copper Alloy Crystals. (Elam)

Material	Tensile Strength lbs./sq in	Elongation %
Copper	17,000	52
Brass I—29% Zn	18,000	72
Brass II—22% Zn	23,500	159
Brass III—25% Zn	23,000	168
Cu-Al I—5% Al	28,000	138
Cu-Al III—5% Al	23,500	110
Cu-Al IV—5% Al	24,000	120

Anisotropy in single crystals. It has not been specifically mentioned but it is clear from the tensile tests of single crystals that they show marked anisotropy. This is entirely consistent with the structure, even though it be cubic, and has been represented formally by means of solid models in which the radius vectors give the values of the yield point and tensile strength in the direction of the vector (52). This is true of even the modulus of elasticity which has been determined for both a FCC metal, copper, (53), and a BCC metal, iron (54). The modulus of iron, E, varies from 18,900,000 pounds per square inch in the [1 0 0] direction to 40,600,000 pounds per square inch in the [1 1 1] direction, as compared to the mean value of 27,000,000 to 30,000,000 pounds per square inch for fine-grained iron. The three major lattice directions of the cubic lattice, [1 0 0], [1 1 0], and [1 1 1] are particularly evident on the space models of the various properties. The non-cubic metals show much more pronounced anisotropic characteristics than the cubic metals do. Some data on selected physical properties of such metals are reproduced in Table IV.

The anisotropic character of single crystals is also brought out by their etching characteristics, particularly the type which develops etching pits. A beautiful example of this has been recently described by Thompson for single crystals of copper, which reproduces the beautiful effects on single crystal spheres of the German workers (55). Likewise the pressure figures, sometimes called Chladni figures, which form when a point or ball is pressed into the surface, reflect the directional character of the surface indented.

Physical Properties of Metallic Single Crystals

Both mechanical and physical properties of the fine-grained metals are statistical means of the properties of the individual grains

in random orientation, plus the intangible effect of the grain boundaries. In technical practice they are, of course, the properties that are of significance, but they do not give proper and adequate criteria of the metallic state. By way of illustration the modulus of iron E, is taken as about 30,000,000 pounds per square inch but for theoretical considerations involving the forces of the iron lattice we should have to utilize the correct physical constants, or the modulii in the principle lattice directions. It is on that account that a summary is given here of some of the physical properties of single crystals. These properties are listed in Table IV in the two directions, parallel to and perpendicular to the major lattice axis.

Differences between the mono- and polycrystalline states. It is not alone by showing marked anisotropic behavior that the single crystal state differs from the polycrystalline state; a few striking differences in behavior have also been observed. The elastic after-affect, which is the gradual return to a stable length of a bar which has been elastically (or plastically) strained and then unloaded, is a well known effect in the common fine-grained metals. It was found by von Wartenberg that single crystals of tungsten and zinc do not show this effect; instead they revert to the stable length immediately whereas the polycrystalline metals require hours (56). This observation has been confirmed by others (57). When this is considered in conjunction with the low measured values of the limit of elastic behavior (and the values cited here are on the high side) we must agree with von Wartenberg that elastic deformations of polycrystalline metals involve plastic strains of at least some of the grains. The elastic after-effect would be caused by the recovery forces exerted by the elastically strained grains on those which were plastically strained, ultimately to bring the test bar back to its initial length.

As further evidence of differences between the two states, we have the absence of "yielding" in single crystal iron, already mentioned. Ziegler reported a large difference in magnetic permeability between single crystal and polycrystalline iron (31), while Geiss and van Liempt found that cold drawing single crystal tungsten produces a much smaller change in the temperature coefficient of electrical resistance than it does with polycrystalline tungsten (58). The striking observation has also been made that single crystal silicon steel does not change in density when cold deformed. This was first reported by O'Neill who found the maximum density change to be 0.07 per cent, or from 7.752 initially, to 7.747 and back to 7.754 finally. In

a tensile test of a single crystal of the same material, the density again remained constant at 7.747 (59). While this relates to silicon iron with 1.81 per cent silicon, Honda and Yamada found the same thing with pure iron. The density fell from 7.870 to 7.860, or by 0.13 per cent as the result of strain, though the Rockwell hardness number increased from 34 to 56 at the same time (60). Pfeil reported an exceedingly small change in the density of iron crystals, or from 7.852 to 7.848 at 71 per cent compression. This amounts to about 0.05 per cent change. Another crystal showed a slight drop at first and then a slight increase with further compression. (30, first paper). This same slight change has also been reported by Sachs and co-workers for single crystals of aluminum and alpha brass (61). The brass crystals showed no density change during single slip but a marked drop occurred with double slip.

Lesser Lattice Changes During Deformation

We have now discussed the major or gross features of the plastic deformation of single crystals. The account is of the form of a report of experimental findings, the accuracy of which depends simply on the accuracy of the means used to study the phenomena involved. On taking up some of the minor lattice changes which accompany deformation, we enter a field which deals with evidence which is much less direct and whose interpretation is by no means as secure. Nor is the agreement of different observers as close here, for with the same observations, different observers have not agreed on the nature of the lattice changes. In spite of this, our interest is no less keen for we are dealing with the particular lattice effects which are held to be responsible for "work hardening," or the strengthening effect of plastic deformation.

One of the first of the lesser lattice changes to receive attention was the distortion of the blocks during plastic deformation. The experimental work of Rosenhain on this point was so skilfully done and is of such historical significance that it requires mention at this place. This is particularly so because he was the first to produce direct evidence of the slip of the "crystalline elements" or blocks over each other. Mügge in Germany (64) and Ewing and Rosenhain in England (65) had already shown that deformation produces slip bands and by direct observation it was clear that the displacement subdivided the grains into numerous thin blocks, though the behavior

of the blocks and the nature of the slip bands themselves was not made clear. It was in attacking this problem that Rosenhain displayed the rare vision and skill in experimental technique for which he subsequently became famous. After utilizing specially developed microscopical methods for observing minute alterations in the shape of the blocks, with results which we shall not detail here, he developed a procedure for studying the precise contour of the metal surface after straining. An iron sample with a well defined system of slip bands was copper plated with a fairly heavy and adherent deposit. By carefully polishing a cross section through this plated surface and examining the edge of the initial section at a high magnification he obtained direct evidence that the original strained surface was stepped or serrated. This was deformation by block movement along slip planes and when considered along with his negative evidence on distortion of the blocks, it appeared that plastic deformation occurred by simple translational displacement of the "crystalline elements" past each other. This conception was combined with the Beilby amorphous metal hypothesis, which had just previously been announced, to account for both deformation and work hardening (66). So appealing was this explanation that it remained the basic thought in this field for many years.

Returning to the absence of distortion of the blocks during deformation, this same position was later adopted by Tammann in his "translation hypothesis" (62), though instead of combining it with the amorphous metal hypothesis to account for work hardening, he utilized the "orientation" principle. While these two authorities were in close agreement on the deformation process, it must not be assumed that no other views were held on this point. At about this same time, it was observed by Osmond and Fremont in studies of the deformation of single crystals of iron that the blocks do become deformed or distorted (82), and this apparent difference in behavior between single crystal and fine-grained iron raised a question which was not answered. At a later date the idea of distortion of the blocks, or for that matter of the whole grains, was carried much further by Czochralski in his "lattice disruption" theory, or "Verlagerungs Hypothese" (63). This rested mainly upon a comparison of the well developed etch figures in the grains of annealed metals with similar effects in the distorted grains of deformed metals. From this it appeared that the lattice became more and more broken up until it attained a completely shattered or amorphous condition.

The most comprehensive survey of the early experimental work in this field is that of Howe (67) though he deals more with the character of the slip bands proper and with the grosser features of deformation, to the almost complete exclusion of the behavior of the metal in the blocks proper.

While this review of early work at its best is but brief and cursory it shows what the situation was before the advent of modern single crystal work and before the methods of X-ray crystal analysis were available to give positive evidence on many of the important points. Perhaps the outstanding contribution is the finding that the crystal lattice remains in a remarkable state of preservation throughout even severe deformation with little or no change in the mean parameter. On the other hand it is also known that profound changes occur in the lattice which are (presumably) sufficiently effective to produce the strengthening observed and even to lead to complete recrystallization of the metal on subsequent heating. These two facts are now well established and they are of such significance that they must guide, or even dominate, any theory of plastic deformation of metals. With this we may proceed to a closer examination of the more recent work which relates to the lesser lattice changes.

In the interests of ascertaining the correct nature of these minor lattice movements, it has seemed necessary to check the extent to which the terrace effect is produced. The first work of Mark, Polanyi, and Schmid (10) suggests that zinc crystals deform without distortion of the blocks and this idea is propagated in the literature by schematic representations of the shape of deformed crystals. An example of this is reproduced in Fig. 2. To check this point the writer made a particular study of the surface of deformed zinc crystals which showed pronounced slip bands. Taking into account the extent of the elongation or the angle of tilt, and the thickness of the white blocks, a simple consideration of the geometry of the slip movement showed that the surface should show terraces of the same order as the thickness of the blocks. Inasmuch as the blocks were visible at reasonably low magnifications the serrations would be readily observed. Much to the surprise of the writer there was no sign of the terrace which it was presumed must be there. Continuing this study we sectioned the crystal correctly to show the terrace on edge, but again the evidence was entirely negative. This work was done with O. E. Romig and has been reported by the writer (68). A schematic representation of the slip movement would

be closer to the truth if no serration were shown. Obviously this discrepancy must be resolved before the lattice movements of the blocks can be properly understood.

A second feature of the microstructure that is not thoroughly understood is the thickness of the two units of the deformed crystal, —the blocks proper and the slip bands. Measurements of both depend on the resolving power and magnification of the microscope used for their observation. One cannot doubt, however, that the lattice at the visible slip bands has been affected differently than it has in the blocks for it is only due to some difference that the slip bands become visible. The theoretical significance of these two magnitudes is such that greater attention to their accurate measurement would be fully justified.

While it has been possible with microscopic examination to ascertain many of the important features of plastic deformation, X-ray analysis has disclosed even more significant evidence of lattice changes. One use of X-ray analysis has already been mentioned, but aside from showing changes in lattice orientation, changes in the character of the diffraction effects have made it possible to follow minor alterations in the lattice.

A Laue pattern (using "white light") of the undeformed crystal shows small and sharply defined spots which correspond to reflections of the various wave lengths from the lattice planes. As deformation progresses these spots increase in size and lose some of their definition. The tendency is to form striations or fourth order diffraction bands, an effect which is known as "asterism." Diffraction patterns with monochromatic radiation likewise show effects of plastic deformation by a broadening of the spots and by the gradual disappearance of the K_α doublet, though without showing any material change in the lattice parameter. A closer examination of these effects would require a detailed study of X-ray methods and would go beyond the limits of the present discussion; on this account the interested reader is referred to the literature on X-ray analysis (69). Here we must be content with a brief summary of the findings.

By following the changes of individual spots of a Laue pattern it is brought out that slip in one direction produces a greater change in the lattice in that direction than in other directions. The effect is similar to that obtained when a mica plate is bent about a digonal axis in the basal plane. Some of the spots are unaffected while others spread out into bands (70). By analyzing patterns from

stretched single crystals in this light it appears that the lattice becomes bent about an axis which lies in the slip plane and at right angles to the direction of slip. In other words it appears that the lattice becomes uniquely strained by plastic deformation while further considerations, which will be mentioned subsequently, show that the lattice cannot be acting as a whole but as more or less individual units. These "fragments" are curved in a crystallographically

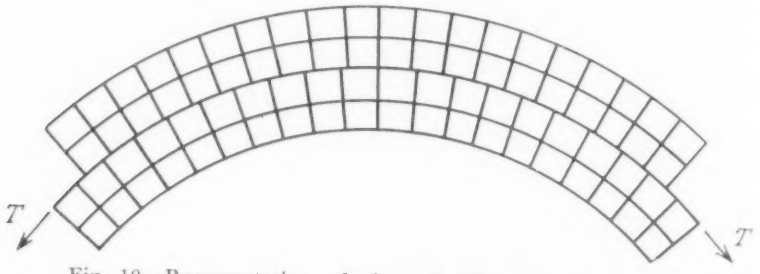

Fig. 10—Representation of Curved "Fragments" in "Flexural Slip." (M. Polanyi, Ref. 71).

defined manner while the thickness and the degree of curvature of the fragments must be consistent with the change in spacing which would be shown in the pattern. This is the conception of "flexural slip" of Polanyi ("Biegegleitung") which is in best accordance with the X-ray findings (71). This scheme is shown diagrammatically by means of Fig. 10.

One point that comes at once to mind is the possible identity of the "blocks" seen under the microscope and the "fragments" of X-ray analysis. Schiebold mentions a figure for the thickness of the blocks which is of the order of 2×10^{-5} centimeters (69) and this is confirmed by similar calculations of Goucher (39). On analyzing his tungsten crystals, Goucher concluded that the lattice fragments must be arranged on curved surfaces to give the broadening of the X-ray lines. On calculating the radius of curvature he found it to be about 9.2 millimeters and he concluded further that the fragments could not be thicker than 2×10^{-5} centimeters to be consistent with a change in the lattice parameter of no more than 0.025 per cent. A check of these magnitudes was secured by following the effect of strain on the resolution of the Cu Kα doublet by use of the Debeye theory. It should be useful to be able to compare this magnitude with the visible block thickness. The writer knows of no such measurements which could be used with confidence, for reasons stated above. There appears to be an opportunity here for someone like

Dr. Lucas to make accurate determinations of the thickness of the blocks at sufficiently high magnification and with a system of sufficiently high resolving power to give the necessary data. Of course, one can make such calculations from photographs of slip bands and in one such case of the writer the average block thickness came of the order of 2.6×10^{-4} centimeters (68).

Summarizing we may say that the evidence affords a reasonably satisfactory analysis of the effects of plastic deformation on the lattice structure of single crystals. First of all the X-ray diffraction pattern shows that the structure is remarkably well preserved even after, and in spite of, severe deformation. This includes the exceptional deformation represented by the after-elongation thread of the zinc crystal which is sufficiently drastic to increase the strength by 50 fold. It is also true that definite changes in the character of the lattice planes are shown by alterations in the diffraction effects so that the end result must be a modified single crystal structure. The effects are such that any motions along the slip planes must occur in approximately whole numbers of the unit lattice distances in the slip direction, and at the same time such that the planes no longer remain "smooth." It is a necessary conclusion that the lattice becomes fragmented though a close description of this condition cannot be given.

Cause of the Increase in Strength

Many are the theories of work hardening or strengthening which have been advanced. Though most of these relate to polycrystalline metals they would also be applicable to, and would have to be consistent with, the behavior of the single crystal. No other field of physical metallurgy has produced so many theories or hypotheses, except the hardening of steel. In the latter case the fundamental data which guide the formulation of a theory were relatively sparse and some of the theories proposed were but mere conjectures. In the present case the conceptions of deformation by slip and twinning were early made clear by Mügge and by Ewing and Rosenhain, and it was further known that distorted transparent crystals showed altered diffraction effects with polarized light. Though this and later work provides a splendid guide, we are still without a theory which properly accounts for both deformation and strengthening. The situation is much the same as it was in 1930 when Schmid, one of the most ardent students of single crystals, stated that while the def-

ormation process of many metals could be correctly stated, it was not known why crystals failed at such low values of the resolved shear stress nor could a quantitative theory for strengthening be set up (*72*).

Jeffries and Archer at one time set up the inquiry,—why are metals so soft? (*73*). The strength or resistance to rupture can be understood on the basis of the force required to separate a metal along some plane and thus overcome the ultimate cohesive forces of the lattice; but why do metals begin to deform plastically at so much lower stresses? They were quite correct in raising this pertinent question but it remained for work with single crystals to show how low such stresses actually are. Thus two cardinal points here are the early failure of the crystal by slip and the subsequent strengthening by deformation of the order of 50 to 1, or possibly even 1000 to 1 if we use recent data for the proportional limit.

It appears that a general theory of the metallic state is in the making, and in the recent book by Hume-Rothery a splendid discussion is given of the present state of this branch of science (*74*). This work is by no means completed and still further from solution is this particular question of the lattice forces which resist the slipping of one plane over another. Hence it is that we do not know whether to assume that the low shear stresses are normal to the metal lattice or whether they are due to imperfection or to some departure from normal. It does not help to use the cohesion forces of the lattice, as has been done for ionic crystals, for they would apply to the resistance of the metal to simple cleavage fracture with no prior deformation. The success which this calculation has had with the ionic lattice checking theory and experiment suggests what may some day be accomplished with the metal lattice. At present we must be satisfied with the assumption that slip along "smooth" planes is easily accomplished.

At this point it becomes apparent that the character of the block movement is important. If it gives the terrace effect, the low yield point applies to the bonds between the blocks, while during deformation the metal of the block would not deform materially and hence would not harden. Initially the metal of the block must have been stronger by many fold than the metal of the potential slip band. If the block movement produces no terrace effect, the deformation must produce appreciable slip in the block with corresponding strengthening. This is consistent with the rapid strengthening of the inactive

planes which has been found by experiment. The first type of movement, as understood at present, is much less satisfactory here.

In the interests of analyzing the nature of this first slip which occurs at low shear stresses we may well consider the conceptions of "thermal" and "athermal" deformation. The former was introduced by Becker to account for the flow of both amorphous materials and metals (76). The application of a low stress to an amorphous body produces a slow deformation or flow. Thermal flow is due to the superposition of elastic strains on the normal thermal vibrations and this produces a change in position of the atoms in their attempt to free themselves of the imposed stress. An attempt at rapid displacement of the atoms by an impact builds up high stresses, which may even exceed the cleavage strength. It seems entirely possible that a metal crystal may respond in the same way when a shearing stress is applied along its slip planes. A mechanism of this type must obviously be affected by temperature in the sense that increasing the thermal vibration of the atoms would require a smaller external stress to produce deformation. We may also expect it to be affected by the rate of applying the stress. Furthermore a load sufficient to produce deformation must also tend to produce subsequent flow by the reason of change in position of the atoms. In the case of the metal lattice such flow would be opposed by the strengthening effect, while the tendency of the crystal to anneal or to recover its initial condition with smooth planes would assist it. This feature is well illustrated by Schmid and Wassermann who showed that hard drawn copper flows faster at 250 degrees Cent. (480 degrees Fahr.) than annealed copper does under the same load. This simulates the flow of amorphous bodies and according to the Becker theory would be due to the same effect,—but there is no suggestion that the hard drawn copper is amorphous (77).

It has been shown that the ductile single crystals of zinc and cadmium retain considerable ability to deform at a temperature of only 20 degrees Kelvin (43) or even 1.7 degrees Kelvin (44). Presumably these crystals would remain ductile at absolute zero. This has led Polanyi and Schmid to assume a second type of deformation, "athermal" deformation, as a characteristic of the lattice and independent of the thermal vibration of the atoms (78). Pure athermal deformation would commence at a stress which is independent of the rate of applying the stress and of the temperature. Observations of single crystal behavior at low temperatures supported their hypothesis

though they could advance no theoretical explanation for it. In this light the ordinary plastic behavior of metals would be due to part thermal and part athermal deformation. With increasing temperature, the effect of "crystal recovery" would become more and more superimposed on the effect of athermal deformation. At room temperature zinc and cadmium would show flow plus strengthening while the low rate of strengthening would be due to recovery. The cubic metals would behave in about the same fashion at elevated temperatures.

With deformation started, why does resistance to further deformation increase? The original work on the slip process and twinning does not touch this point but subsequently many opinions have been expressed on it. Few of these ideas can be proved or disproved; none can be used quantitatively or to predict behavior. One of these ideas has had the strong appeal of simplicity and has been given such wide publicity that we may well inquire here into its plausibility in view of the work on single crystals. I am referring to Beilby's amorphous metal hypothesis (86). According to this hypothesis deformation on a slip plane generates amorphous metal along that plane which is harder and stronger than the crystalline phase; subsequent slip is made more difficult and the metal is harder. In 1927 the writer showed by the limited work on zinc crystals that Beilby's hypothesis was untenable in view of the rotation of the central portion of the crystal about the hexagonal axis (68), and the broader field covered here fortifies that position. In fact, our present test is definitely negative. During slip on the initial set of slip planes, the resistance to slip along the active plane increases; but resistance to both slip and cleavage increases at a still greater rate on planes which cross at an angle. By no conceivable mechanism could amorphous metal form along the latent slip plane to account for its greater shearing stress during bilateral slip. The effect on the resistance to rupture, as was discussed particularly for zinc, is equally as positive and even more striking. There have always been grave doubts of the validity of the Beilby theory but quantitative single crystal work rules it out definitely. Its principal deficiency as a working hypothesis has been the complete lack of proof of the presence of the amorphous metal that was assumed. The Czochralski lattice disruption theory, which is an extrapolation of major proportions of the Beilby theory, is accounted for by the proof that the

single crystal state is so well preserved throughout even drastic deformation.

Of much greater utility are generalizations which coordinate previously unrelated phenomena and establish a common viewpoint from which they may be examined. The first comprehensive principle in this field was that of Ludwik who conceived the fruitful idea that hardness or resistance to deformation was affected chiefly by the perfection of the planes which accommodate slip. Anything which vitiated that perfection would obstruct slip and make the metal harder or stronger. Such obstructing agencies might be cold work, alloying, or heat treatment (79). Later it was found that plastic deformation must fragment and crumble the lattice and in the correct manner to obstruct slip and when these findings were formulated by Polanyi into a working hypothesis, it was quite consistent with Ludwik's conception.

A similar generalization was independently arrived at, and at about the same time, by Jeffries and Archer (73). Their well known "slip interference" theory is too well known to require detailing here and it was designed mainly for polycrystalline metals. The cardinal point of this theory is contained in its appropriate name. Rosenhain also correlated many aspects of metallic behavior in terms of the condition of the lattice and we may say that the broad principles of these generalizations are invaluable in the guidance they give to thought along these lines. The major difficulty is encountered when an attempt is made to predict behavior quantitatively, and this becomes particularly apparent when a simple mechanical picture is utilized. Just as classical mechanics had to give way to wave mechanics for problems of electron movements, it is clear that the mechanical analogy will be incapable of clarifying the problem of plastic deformation.

Goucher may have been quite correct when he concluded that the strengthening of his tungsten crystals was due to "residual elastic strains produced by the deformation process rather than the mere fact of crystal break up," but what is the next step in the theory? We may sense in all of this the trend in thought that whatever it is in the deformation process that alters the physical character of the lattice planes, slip planes or others, introduces a mechanical, atomic, or electronic disturbance which gives a structure of a higher degree of stability when acted upon by deforming or rupturing forces.

If it were attempted here to discuss the "ultimate" cause of the

changes in properties, we would enter a field of almost pure speculation. On that account the author will simply indicate three main lines of thought.

One assumption has its basis in the minor lattice changes which are observed by X-ray analysis; or in other words, the idea of flexural slip can be utilized to account for strengthening. The lattice planes are rumpled across the direction of slip to make slip more difficult in that direction while the disregistry at the interfaces of the fragments makes slip more difficult on the latent slip planes. By means of mathematical treatment Dehlinger has sought to show that rows of atoms can be displaced to new positions and leave a nearly stable configuration which remains after the load is removed and leaves the lattice in a strained condition (87). This line of thought requires the lattice to change with deformation to introduce the strengthening effect by mechanical means.

A second line of reasoning is based on the assumption that the crystal originally contains defects in the form of discontinuities in the lattice or internal notches or defects. It is also assumed here that the inherent strength of the lattice (cohesive forces) is not, and cannot be, altered by plastic deformation; the result of plastic deformation is to lower the effectiveness of such defects. The basis of this idea is the high calculated theoretical strength of a crystal as compared to the measured strength.

The third idea rests on the assumption that plastic deformation alters the character not alone of the lattice but of the atoms (electron orbits) themselves. It is significant here that properties other than the mechanical properties are altered by plastic deformation, some of which are definitely associated with the electrons. This would indicate that the lattice and the cohesive forces thereof are altered. Inasmuch as there is a sharp schism between these three lines of thought, it is possible that we may soon be able at least to discard two of them. It should be helpful in this connection to consider in detail the chemical and electrical properties of the metals, as some have already done, and decide the main point at issue—if the properties of the lattice are actually altered by deformation. It seems to me that the work of Kuntze on the cleavage strength of metals speaks powerfully for the third idea and the increase due to cold working which he found is confirmed by investigations of single crystals (75). Another powerful argument is found in the failure of the Cauchy relations for the metals, for it follows logically from that circumstance

that the atoms or ions alone are not the units of structure of the lattice, as is assumed for those relations, and hence it appears that the lattice behavior depends on the electrons too. (See 81).

In closing, the writer realizes that much more could be said about this important and interesting problem of metallic behavior. In the present resume the experimental side has been emphasized because the findings can be used by the metallurgist without regard to the reasons back of the behavior.

Table IV

Physical Constants of Certain Metals

|| and ⊥ to Princ. Axis	Elast. Mod. E—Psi	Lin. Compressibility cm^2 Kg (1)	Spec. Elec. Res. cm—ohm—0°C	Lin. Coef. of Exp. 20°C	Observer
Zinc	||	5.15×10^6	15.8×10^{-7}	5.83×10^{-6}	63.0×10^{-6}	Grüneisen & Goens (2)
Zinc	⊥	17.05	1.4	3.59	14.2	
Zinc	||	5.50	12.98	5.66	57.4	Bridgman (3)
Zinc	⊥	17.58	1.95	5.45	12.6	
Cadmium	||	4.10	16.6	7.79	5.25	Grüneisen & Goens
Cadmium	⊥	11.80	1.5	6.54	20.2	
Cadmium	||	3.92	....	7.65	....	Bridgman
Cadmium	⊥	11.24	....	6.25	....	
Tin	||	12.30	6.72	13.1	30.5	Bridgman
Tin	⊥	7.83	6.02	9.1	15.45	
Antimony	||	4.29	16.48	31.8	15.56	Bridgman
Antimony	⊥	8.19	5.26	38.6	7.96	
Bismuth	||	....	....	143	....	Borelius & Lindh (4)
Bismuth	⊥	....	....	107	....	
Bismuth	||	5.01	15.92	127	13.96	Bridgman
Bismuth	⊥	5.39	6.62	100	10.36	
Tellurium	||	6.18	—4.14	56×10^{-3}	—1.6	Bridgman
Tellurium	⊥	2.97	27.48	154	27.2	

(1) The second degree term is omitted from the compressibility.

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CREEP CHARACTERISTICS OF METALS

By C. L. Clark and A. E. White

Abstract

This paper, after briefly defining the term creep and offering an explanation of the creep phenomenon, discusses the various factors which may influence high temperature load-carrying ability. Among those considered are chemical composition, structural uniformity, heat treatment, method of manufacture, grain size, previous deformation, equicohesive temperature, and structural stability.

It shows that creep is a sensitive property which is influenced to a greater degree by variations in the past history of the material than the other physical properties ordinarily determined. Care should, therefore, be exercised in applying creep data obtained from one material to another unless the complete past history of both is known to be the same.

IN a symposium devoted to the plastic properties of metals, attention should certainly be given to the subject of creep. Not only is this distinctly a plastic phenomenon but it is also one that has been of great concern during the past few years to designers and users of high temperature equipment.

This subject has been broadly investigated and, as a result, hundreds of articles have appeared in the literature since 1920. It is, of course, impossible for a single paper of this type to fully review all of this work as such a reference would of necessity be very extensive. The present paper will, therefore, be confined to a definition of creep, to a brief explanation of the creep phenomenon, and to a brief discussion of the various factors which may influence the creep resistance.

Definition of Creep

Steels, as well as most metals and alloys, can be considered for all practical purposes as possessing elastic properties at room tem-

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perature. Accordingly, if a stress below the proportional limit is applied, such as stress X of Fig. 1, an elastic deformation will occur immediately upon the application of the load and regardless of the length of time the stress is applied the deformation will remain con-

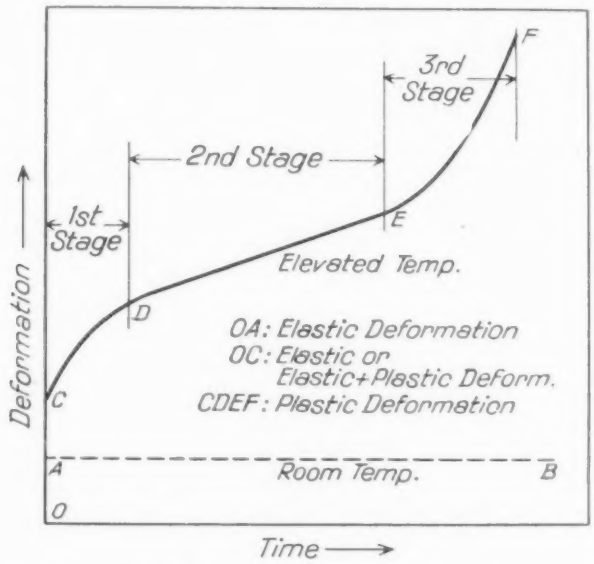

Fig. 1—Sketch Defining the Meaning of the Term "Creep."

stant within present measurable limits. In other words, the deformation characteristics under this condition will be represented by the curve OAB.

At elevated temperatures a deformation OC likewise occurs upon the application of the stress Y and this may be entirely elastic or elastic plus plastic, depending upon the material, the temperature, and the stress. Under suitable stress-temperature conditions, however, the deformation increases as the time of load application is extended, the deformation following the curve CDEF. This increase in deformation with time under a constant stress is defined as creep.

As indicated in Fig. 1, the complete creep curve consists of three stages. In the first stage creep occurs at a decreasing creep rate; in the second, at an approximately constant creep rate; and in the third, at an increasing creep rate. In the majority of tests conducted, however, the stress and temperature conditions are selected so that the third stage is not entered. While too broad generaliza-

tions are not permissible it is generally found that the duration of each stage tends to vary inversely with the stress.

Explanation of Creep

The question then arises as to why metals creep at elevated temperatures and what factors are responsible for the change from the elastic to the plastic state. No explanation has yet been advanced which meets with the approval of all those most familiar with this phenomenon.

The conception of creep which appears to be most in harmony with the observed behavior of metals subjected to stress at elevated temperatures may be summarized as follows: At elevated temperatures at or below the equicohesive temperature, which is the lowest temperature of recrystallization, the deformation or creep resulting from the application of a given stress over a given time period is a running balance between the yielding of the material and the strain-hardening caused by such yielding. At or below this range strain-hardening tends to predominate and continuous measurable creep will not occur unless the stresses are sufficiently great to overcome the resistance caused by strain-hardening. In fact, the yielding of steel at room temperature when the proportional limit is exceeded is in reality a creep phenomenon but unless the stress is sufficiently great the measurable deformation soon stops as a result of the marked rate of strain-hardening.

At temperatures above the equicohesive temperature, however, the recrystallization rate exceeds the strain-hardening rate and creep will proceed under the lowest stresses. In fact, it is believed that creep could be detected at these higher temperatures under any stress whatsoever if sufficiently sensitive apparatus were available.

It is also believed that the mechanism of deformation changes as the testing temperature is increased. At temperatures below the equicohesive temperature deformation apparently occurs largely within the crystals themselves and since these crystals are perfect elastic bodies the metal exhibits elastic properties. At temperatures above the equicohesive temperature, however, it is felt that if sufficient time is allowed the deformation occurs through the movement of the crystals themselves, that is, within the materials surrounding the grains. Regardless of the nature of this material and whether it is amorphous or severely-strained crystals, it is generally

agreed that it is not elastic. Accordingly, metals above this critical temperature exhibit plastic rather than elastic properties.

These statements are recognized to be hypotheses rather than proven laws, but as will be shown they do permit reasonably satisfactory explanation of the various creep phenomena. It is possible that as additional information is acquired these hypotheses may be revised or changed, but on the basis of present knowledge they afford a plausible explanation of the high temperature behavior of metals.

Factors Influencing Creep Resistance

On the basis of work which has been done, creep is believed to be one of the most sensitive physical properties determined and is, therefore, greatly influenced by many factors which have little or no effect on the other usual physical properties. For example, a series of steels having practically the same tensile properties and hardness at room temperature may show differences of several hundred per cent in their creep resistance at higher temperatures. Likewise, two steels of identical chemical composition, but made in different type melting furnaces, may possess approximately the same room temperature physical properties, yet show great differences in their creep strength.

Owing to the complexity of the subject information is not available concerning all the possible variables and, in fact, probably all of them are not as yet even known. Information is, however, available with respect to many of them and among those to be considered in the present discussion are chemical composition, structural uniformity, method of manufacture, heat treatment, grain size, previous deformation, equicohesive temperature, and structural stability.

Certain of these factors are, to some extent, overlapping. For example, the heat treatment will influence the actual grain size and often the structural stability. Likewise composition may influence the structural stability and the equicohesive temperature. Nevertheless, it is believed that each is worthy and capable of individual consideration.

It is likewise known that variations in testing procedure, such as duration of test, method of loading or heating, and sensitivity of extensometer system, will influence the observed creep characteristics. The present discussion will, however, be limited to those factors pertaining to the material itself.

Chemical Composition—For many years it was believed that all low-alloyed steels of the pearlitic type possessed the same creep characteristics at temperatures of 1000 degrees Fahr. or greater. If superior creep resistance were required it would, therefore, be necessary to resort to highly alloyed austenitic steels, such as the 18 per cent chromium, 8 per cent nickel type, or the ferritic steels of the 18 per cent chromium type.

It is now generally agreed, however, that a wide range in the creep characteristics may be obtained at temperatures up to 1100 to 1400 degrees Fahr. with pearlitic steels depending upon their composition. The magnitude of the range in the creep strength of certain steels of this type is illustrated by the values given in Tables I and II. Table I considers low carbon alloy steels containing 0.10 to 0.15 per cent carbon while in Table II the carbon content varies from 0.35 to 0.45 per cent. In Table I all other variables, such as method of manufacture, heat treatment, and inherent grain size, were kept constant while in Table II the heat treatment was varied in order to produce a constant hardness of 285 Brinell. Referring to Table I it is observed that the stress for a creep rate of 0.10 per cent per 1000 hours varies at 1000 degrees Fahr. for the low carbon alloys considered from 5750 to 24,000 pounds. Likewise, with the higher carbon materials, Table II, the range at 1000 degrees Fahr. is from 780 to 24,000 pounds.

The desired operating temperature determines the most suitable composition. With temperatures below the lowest temperature of recrystallization (equicohesive temperature) the creep resistance may be increased by the addition of elements which largely enter into solid solution with the ferrite, such as nickel and manganese, as well as by the addition of carbide-forming elements, such as chromium, tungsten, molybdenum and vanadium. It is believed that the effectiveness of these latter elements is due to their carbides being located within the crystals as well as in the grain boundaries; they thus exert a keying action along the crystallographic planes and thereby decrease the tendency for slippage or creep.

At temperatures above the lowest temperature of recrystallization the elements most effective in increasing the creep resistance of pearlitic-type steels are those forming carbides. As it is felt that creep occurs at these temperatures largely as a result of the movement of crystals with respect to one another, rather than by slippage within the crystals themselves, the effectiveness of the carbide-

Table I

Influence of Chemical Composition

Creep Characteristics of Pearlitic Tube Steels at Designated Temperatures

All Specimens were Annealed from 1550 Degrees Fahr., were from Electric Furnace Heats and were Coarse-Grained

Material	Type Chemical Composition				Temperature Degrees Fahr.	Stress for Designated Rate of Creep Per Cent per 1000 Hours		
	C	Si	Cr	Mo		0.01	0.10	1.0
1015	.15	.23	...	...	800	18,500	26,800	38,500
C-Mo	.16	.23	...	.42	800	15,500	26,000	44,500
Si-Mo	.11	1.35	...	.50	800	18,000	29,500	48,500
Cr-Si-Mo	.07	.72	1.24	.54	800	20,000	29,000	42,500
Si-Cr-Mo	.11	.78	2.50	.50	800	22,000	27,000	33,000
Cr-Mo	.10	.18	5.09	.55	800	14,250	22,000	34,000
1015	.15	.23	...	...	900	12,800	16,900	22,100
C-Mo	.15	.23	...	...	1000	2,700	5,750	12,100
Si-Mo	.16	.23	...	.42	1000	10,700	17,800	29,200
Cr-Si-Mo	.11	1.35	...	.50	1000	5,000	9,200	17,000
Si-Cr-Mo	.07	.72	1.24	.54	1000	15,000	24,000	37,500
Cr-Mo	.11	.78	2.50	.50	1000	6,100	9,300	13,500
1015	.10	.18	5.09	.55	1000	7,000	10,250	12,900
Cr-Si-Mo	.15	.23	...	...	1100	840	1,800	3,850
C-Mo (Cal)	.07	.72	1.24	.54	1100	4,300	6,800	10,800
1015	.16	.23	...	.42	1100	2,700	7,000	15,000
C-Mo	.15	.23	...	...	1200	290	620	1,300
Si-Mo	.16	.23	...	.42	1200	480	2,000	4,050
Cr-Si-Mo	.11	1.35	...	.50	1200	720	1,850	3,250
Si-Cr-Mo	.07	.72	1.24	.54	1200	1,950	3,950	8,100
Cr-Mo	.11	.78	2.50	.50	1200	1,000	2,325	5,350
C-Mo (Cal)	.10	.18	5.09	.55	1200	900	2,500	5,250
Cr-Si-Mo (Cal)	.15	...	...	1.00	1200	1,100	2,600	6,200
C-Mo (Cal)	.07	.72	1.24	.54	1300	700	1,800	3,900
C-Mo (Cal)	.16	.23	...	.42	1300	210	840	1,900
Cr-Si-Mo (Cal)	.15	...	...	1.00	1300	840	2,000	4,800
C-Mo (Cal)	.07	.72	1.24	.54	1400	250	560	1,100
	.15	...	...	1.00	1400	380	750	1,475

Cal = Calorized Specimens.

Note: Majority of Creep Tests of 1000 Hours Duration.

Table II
Influence of Chemical Composition
Creep Characteristics at Elevated Temperatures of Alloy Bolt Steels

All Specimens Heat Treated to Hardness of 285 Brinell

Material	Chemical Composition							Heat Treatment Degrees Fahr.	Temperature Deg. Fahr.	Stress for Designated Rate of Creep Per Cent per 1000 Hours		
	C	Si	Cr	Mo	Ni	W	V			0.01	0.10	1.0
Cr-W-Si-V	.45	.77	1.40	...	...	.98	.25	N. 1725, D. 1300	800	48,000	68,000	96,000
Cr-W-Si-V	.45	.77	1.40	...	...	.98	.25	N. 1725, D. 1300	1000	13,750	24,000	42,500
Cr-W-Si-V	.45	.77	1.40	...	...	.98	.25	N. 1725, D. 1300	1200	540	1,200	2,650
Cr-Si-Mo	.48	.62	1.20	.52	...	...	...	N. 1725, D. 1180	800	47,000	92,000	.....
Cr-Si-Mo	.48	.62	1.20	.52	...	...	...	N. 1725, D. 1180	1000	7,800	23,000	68,000
Cr-Si-Mo	.48	.62	1.20	.52	...	...	...	N. 1725, D. 1180	1200	215	900	3,900
4140	.40	.25	.92	.18	...	...	...	O. Q. 1550, D. 1140	1000	3,700	7,400	15,000
Cr-Ni-Mo	.39	.27	.72	.32	1.74	...	...	O. Q. 1525, D. 1125	1000	.....	780	13,500
3140	.36	.26	.56	...	1.33	...	...	O. Q. 1550, D. 1000	1000	190	1,025	5,200
6140	.40	.21	1.01	...	...	...	.17	O. Q. 1600, D. 1200	1000	1,400	3,900	8,800

Note: Creep Tests of 500 Hours Duration.

forming elements in reducing creep is believed to be due to the fact that a large number of carbide particles are located at the grain boundaries and thus restrain the relative movement of the crystals.

Certain alloying elements may also improve the creep resistance at a given temperature by increasing the lowest temperature of recrystallization. Since creep and the lowest temperature of recrystallization or equicohesive temperature appear to be closely connected any element which increases this temperature will also increase the creep resistance at temperatures above it. Very little is at present known as to the effect of the addition of alloying elements upon the lowest temperature of recrystallization, but this is a field which is now receiving attention.

Influence of Typical Elements—The belief was formerly held that for certain elements at least the creep resistance was directly proportional to the amount of alloy addition. That is, the greater the alloy content the greater would be the creep strength. This is now known to be untrue and, in fact, as shown by certain of the results of Table I, the converse may often be correct. In any type series of pearlitic alloy steel evidence indicates that certain critical amounts of each of the elements must be present to produce the maximum creep resistance. It is, therefore, somewhat difficult to consider each element separately unless it is realized that only trends can be developed.

A common way of increasing the room temperature strength of a steel is by raising the carbon content and it was only natural that the same procedure should be applied to high temperature material. Many conflicting statements can still be found in the literature, however, as to the influence of this element on the creep strength. Tapsell[1] has stated that increased carbon content is beneficial in plain carbon steels at lower temperatures but that it is without appreciable influence at temperatures in excess of 840 degrees Fahr. Data provided by Norton[2] indicate that at 1000 degrees Fahr. the creep strength is increased as the carbon content is raised from 0.08 to 0.42 per cent. Grün[3] found that at temperatures of 750 to 930 degrees Fahr. the creep strength might actually decrease as the carbon exceeded 0.30 per cent, while Juretzek and Sauerwald[4] found the

[1] "Creep of Metals," 1931, p. 215.

[2] "Creep of Steel at High Temperatures," 1929, p. 88.

[3] "The Creep Strength of Steels as Dependent on Alloy and Heat Treatment," *Arch. Eisenhüttenw.*, Vol. 8, 1934, p. 205-11.

[4] "Creep Resistance of Structural Steel and a Simplified Testing Method," *Die Wärme*, Vol. 57, 1934, p. 267-73.

Table III
Influence of Chemical Composition
Influence of Carbon Content on the Creep Resistance of a Plain Carbon and Low Alloyed Steel

Steel	Temperature Degrees Fahr.	Stress for Designated Rate of Creep Per Cent per 1000 Hours		
		0.01	0.10	1.0
1015	800	12,000	17,200	25,500
1040	800	13,000	18,500	27,000
1015	1000	1,800	3,300	6,000
1040	1000	2,800	5,500	9,300
1015	1200	140	540	2,100
1040	1200	140	860	2,650
Si-Cr-Mo (.07C)	800	20,000	29,000	42,500
Si-Cr-Mo (.48C)	800	15,000	24,500	40,000
Si-Cr-Mo (.07C)	1000	13,000	22,500	38,000
Si-Cr-Mo (.48C)	1000	4,100	8,200	15,000
Si-Cr-Mo (.07C)	1200	1,950	3,950	8,100
Si-Cr-Mo (.48C)	1200	1,000	2,200	4,800

Note: Creep Tests of 1000 Hours Duration.

creep strength to increase with increasing carbon content at 750 to 1020 degrees Fahr.

The complete past history of the steels used by these investigators was not always known and it is entirely possible certain of the variations observed might have been due to factors other than the carbon content. The authors[5] investigated a series of plain carbon and alloy steels in which all variables except carbon content were controlled and from their results, given in Table III, it is apparent that in the plain carbon steel the increase in carbon content was beneficial at temperatures up to 1000 degrees Fahr. while in the alloy steel increased carbon content resulted in lower creep strength at all three temperatures considered (800, 1000 and 1200 degrees Fahr.).

On the basis of these facts it appears that too broad statements are not permissible concerning the influence of carbon content. It is believed, however, and supported by the data of Table III, that increased carbon content raises the creep strength when the carbides are in a lamellar form and decreases the creep strength when the carbides are in a spheroidized condition.

Chromium is an element added to the majority of high temperature steels not only for increasing the creep strength but for improving the oxidation and corrosion resistance as well. However, as shown in Fig. 2, the creep strength is not directly proportional to

[5]"Influence of Carbon Content on the High Temperature Properties of Steels," TRANSACTIONS, American Society for Metals, Vol. 23, 1935, p. 995-1015.

the chromium content at any of the three temperatures considered. In fact, the maximum occurs at about 1.25 per cent and decreases as the chromium content is either raised or lowered. This relationship should be considered as applying only for the given carbon content, 0.15 maximum, and the given molybdenum content, 0.50 per cent.

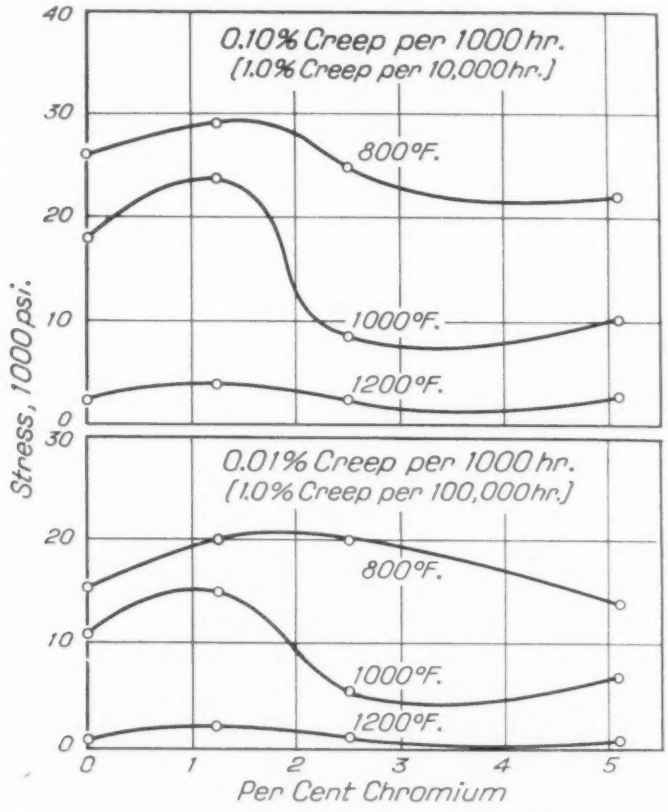

Fig. 2—Influence of Chromium Content on the Creep Resistance of Annealed, Electric-Furnace 0.50 Per Cent Molybdenum Steel. Creep Tests of 1000 Hours Duration.

It is entirely possible that this relationship might be altered if these elements were varied or if additional ones were added.

Molybdenum, tungsten and vanadium are similar to chromium in that they are primarily carbide formers. They likewise increase the creep strength at elevated temperatures. While the data with respect to these elements are not as complete as with chromium it is known that, in the case of molybdenum at least, the creep strength continues to increase as this element is raised from 0.25 to 1.50 per cent. It is entirely possible that a certain critical amount of this element will also give the maximum resistance, but the magnitude of this amount is not as yet known.

Manganese differs from the elements previously considered in that it largely enters into solution in the ferrite. As in the case of chromium, however, definite amounts of this element are required for any given type alloy to produce the maximum creep resistance. This is illustrated by the data of Fig. 3 in which the manganese content of a 0.25 per cent molybdenum steel was varied. It is interesting to note that the maximum creep strength, as with chromium, occurs

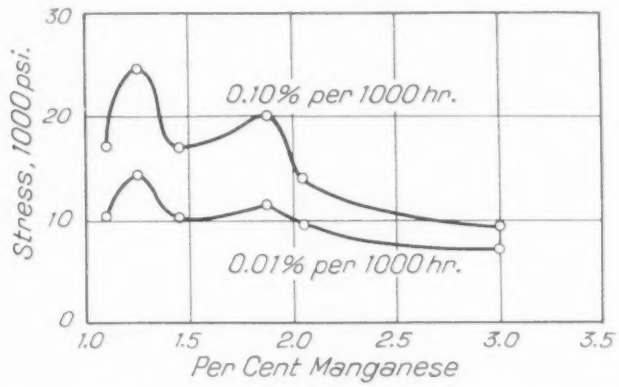

Fig. 3—Influence of Manganese Content on the Creep Resistance at 900 Degrees Fahr. of Electric-Furnace 0.25 Per Cent Molybdenum Steel. Creep Tests of 500 Hours Duration.

at 1.25 per cent. A secondary maximum also occurs at 1.88 per cent but the strength of the higher manganese alloys is considerably lower.

Although data are not available it is probable that elements such as nickel, silicon, aluminum, and possibly copper will have the same general influence on the creep resistance as manganese. In fact, preliminary results indicate that silicon additions in excess of a given amount do lower the creep resistance.

While the preceding comments have applied particularly to low-alloyed steels of a pearlitic type the same general conclusions may also be applied to highly alloyed and austenitic steels. For example, the carbide-forming elements, such as tungsten and molybdenum, are generally added for increasing the creep resistance while other elements, such as silicon or aluminum, are added primarily for improving certain other properties. The general influence of the austenitic-forming alloys, such as nickel and manganese, would, however, show marked discontinuities as the iron changes from the alpha to the gamma form.

Structural Uniformity—Before the influence of the various factors on the creep characteristics was known, disagreements in the

observed creep values of steels of the same type composition and heat treatment were often attributed to lack of structural uniformity. While this factor is important, it should be realized that it is not the only one which may influence the results.

If pronounced chemical segregation occurs and the alloying elements which have been added for increasing the creep strength are not uniformly distributed, it is apparent that the resistance to creep will be largely determined by the strength of the weakest portion of the mass. This chemical segregation is often manifested by banding of the structure. Wyman[6] investigated this phenomenon, using a series of chromium-nickel-molybdenum steels, and found banding to exert a great influence. In fact, at 842 degrees Fahr. (450 degrees Cent.) the values reported ranged from 11,500 to 36,000 pounds per square inch, a difference of over 300 per cent. The fact that the steels were not of exactly the same composition and were secured from different sources might indicate that other factors were also influencing the results. That this is not the case is substantiated by the results obtained when the banding was eliminated through heat treatment.

Structural nonuniformity is also often produced during the service, especially in certain of the austenitic chromium-nickel steels. With these materials carbide precipitation and migration occurs under certain conditions and in the extreme cases each grain may be completely surrounded by a carbide network. Very little information is available in the literature concerning the influence of this change on the creep resistance. Norton[7] did, however, conduct creep tests on 18-8 after four years of service and in the specimens he used carbide precipitation had occurred. His conclusions indicate that the precipitation had no influence on the resulting creep strength.

Under certain operating conditions pearlitic type steels likewise undergo a structural change in that the pearlite changes from a laminated to a spheroidized form. This, however, cannot be considered as producing structural nonuniformity and will be considered in later sections.

Heat Treatment—Heat treatment is widely recognized as having an important effect on the creep resistance of metals although there is not as yet complete agreement as to the most suitable heat treat-

[6] "The Creep of Steels as Influenced by Macrostructure," *Mechanical Engineering*, Vol. 57, No. 10, 1935, p. 625-27.

[7] "Creep Strength of 18-8 After Four Years' Service," *Metal Progress*, Vol. 25, 1934, p. 50.

ment. In general, it may be stated that the heat treament should be such that a stable structure results at the maximum operating temperature. However, as will be discussed later, this is not always possible and, furthermore, there are differences of opinion as to which heat treatment produces the structure of maximum stability.

Many believe that all high temperature steels should be used in the annealed state since they feel that service conditions will ultimately reduce the constituents of the steels to a form in which they would have existed had the material been annealed. Lately there has been a question as to whether annealing results in the greatest stability especially for steels used at operating temperatures of 1000 degrees Fahr. or greater. In many steels the annealing treatment produces lamellar pearlite although, as pointed out by Bailey and Roberts[8] and since verified by many, the pearlite may change during service from a laminated to a spheroidized condition. Certain investigators then advocated that the creep characteristics at least should be determined on spheroidized structures, believing that this represented the condition of maximum stability.

This problem is further complicated by the fact that a spheroidized structure in itself may not be the condition of maximum stability. It has been found that spheroidization, under certain conditions at least, occurs in several steps. In the first, the lamellar pearlite is transformed to a condition commonly described by the term sorbitic pearlite. These carbide particles then increase in size and migrate and if suitable conditions exist the migrations to the grain boundary may be practically complete. There is also a question as to whether even this stage represents the condition of maximum stability.

It appears, therefore, that for operating temperatures of 1000 degrees Fahr. or greater the structure may not be absolutely stable regardless of the heat treatment and that recourse must be had to the one possessing the greatest degree of stability. In general, this can best be accomplished by annealing although it is believed that a normalized and drawn structure will also possess a high degree of stability provided the maximum operating temperature is about 200 degrees Fahr. below the drawing temperature employed. This latter treatment has an advantage in that it will generally produce a greater creep resistance than annealing operation. In this connec-

[8] "Testing of Materials for Service in High Temperature Steam Plant," *Engineering*, Vol. 133, 1932, p. 261-65, p. 295-98.

Table IV

Influence of Heat Treatment on the Creep Characteristics at 1000 Degrees Fahr. of Electric and Open-Hearth 1015 Type Steels

Designation	Heat Treatment Degrees Fahr.	Stress for Designated Rate of Creep Per Cent per 1000 Hours		
		0.01	0.10	1.0
El-A	Hot Rolled	7,000	12,800	15,800
OH-A	Hot Rolled	6,100	6,800	7,800
El-B	N. 1725, D. 1200*	4,900	9,900	12,100
OH-B	N. 1725, D. 1200*	4,000	5,500	7,700
El-C	Ann. 1550	3,000	6,200	12,900
OH-C	Ann. 1550	2,350	4,500	6,200
El-D	N. 1725, D. 1200**	2,750	5,800	8,200
OH-D	N. 1725, D. 1200**	2,600	5,300	7,400

*Drawn 1 hour at 1200 degrees Fahr.

**Drawn 168 hours at 1200 degrees Fahr.

Note: Creep Tests of 1000 Hours Duration.

tion Wyman[6] found that a drawing temperature of 1110 degrees Fahr. (600 degrees Cent.) produced a structure which was stable after many thousands of hours at 840 degrees Fahr. (450 degrees Cent.) but one that was unstable when the testing temperature was in excess of 1022 degrees Fahr. (550 degrees Cent.).

Table IV shows the influence of four different heat treatments on the creep resistance of two carbon steels at 1000 degrees Fahr. These steels were of approximately the same composition but differed in that one was made in the electric arc furnace and the other in the open-hearth furnace. The heat treatments considered were the hot-rolled, annealed, normalized and drawn, and spheroidized. The spheroidized condition was produced by maintaining the specimen at the drawing temperature, 1200 degrees Fahr., for 168 hours.

From the results it is apparent that the hot-rolled steel possesses the maximum creep strength in both cases. The advantages of this heat treatment, however, can be regarded as only temporary for the structure is the most unstable of those examined. If an alloy steel possessing marked air hardening characteristics were being considered, the structural instability would be even more pronounced.

The normalized and drawn specimens possessed the next highest creep resistance. In the case of the electric furnace steel the annealing treatment produced a higher creep resistance than the spheroidizing treatment while with the open-hearth steel the influence of these two heat treatments was reversed. On the basis of these results, therefore, it must be concluded that spheroidizing does not necessarily produce the minimum creep strength. It should further

Table V

Influence of Heat Treatment on the Creep Characteristics of a Si-Cr-Mo* Steel at 800, 1000 and 1200 Degrees Fahr.

Heat Treatment	Temperature Degrees Fahr.	Stress for Designated Rate of Creep Per Cent per 1000 Hours		
		0.01	0.10	1.0
N. D.**	800	47,000	92,000	.....
Annealed	800	19,000	27,000	38,000
N. D.	1000	7,800	23,000	68,000
Annealed	1000	4,100	8,200	15,000
N. D.	1200	215	900	3,900
Annealed	1200	1,000	2,200	4,800

*0.48C, 0.49Mn, 0.62Si, 1.20Cr, 0.52Mo.

**Normalized 1725 degrees Fahr., Drawn 1180 degrees Fahr.

Note: Creep Tests of 500 Hours Duration.

be noted that with these two steels at least the differences between the results obtained from the annealing and spheroidizing treatments were not especially marked, particularly at the two lower creep rates.

The relative merit of annealing and normalizing treatments for producing a high creep resistance is illustrated by the data of Table V. At the two lower temperatures considered, 800 and 1000 degrees Fahr., the normalizing treatment produced the maximum creep resistance while at 1200 degrees Fahr. the annealing treatment gave the superior results. The low resistance of the normalized specimens at 1200 degrees Fahr. is largely due to the fact that the drawing temperature employed was 20 degrees Fahr. below the actual testing temperature. The structure was, therefore, unstable under the given conditions.

Varying certain factors in the heat treatment operation may also influence the creep resistance. For example, annealing or normalizing treatments conducted at temperatures greater than normal will result in a larger actual grain size and this, in turn, will affect the creep resistance. This matter, however, will be considered in connection with the influence of grain size on creep.

Summarizing with respect to the usual types of heat treatment, it may be concluded that in general the creep resistance is greatly influenced by heat treatment and further that at temperatures of 1000 degrees Fahr. or greater the maximum resistance is produced by normalizing provided the drawing temperature is about 200 degrees Fahr. above the test temperature and the lowest resistance by quenching and drawing. The creep resistance of the annealed steels is between those for the other two treatments. This relationship may

Table VI

Influence of Melting Process
Creep Characteristics at Indicated Temperatures of Electric and Open-Hearth 1015 Type Steels (Annealed)

Type Steel	Temperature Degrees Fahr.	Stress for Designated Rate of Creep Per Cent per 1000 Hours		
		0.01	0.10	1.0
Electric	800	18,500	26,800	38,500
Open Hearth	800	12,000	17,200	25,000
Electric	900	12,800	16,900	22,100
Electric	1000	2,700	5,750	12,100
Open Hearth	1000	1,800	3,300	6,000
Electric	1100	850	1,800	3,850
Electric	1200	290	620	1,300
Open Hearth	1200	140	540	2,100

All values based on tests of at least 1000 hours duration.

not hold for all alloy steels but it has been found to be true for the majority of those examined.

Method of Manufacture—Even though two steels have substantially the same chemical composition the creep characteristics may differ according to the manufacturing methods employed. The factors to be considered in this connection are the melting process, the melting practice, and the casting practice. A distinction must be drawn between melting process and melting practice. The melting process refers to the types of furnaces used in making steel and the melting practice to the manner in which the liquid steel is handled in the furnace itself or in the ladle after tapping the furnace.

Melting Process—The two most commonly used furnaces for producing steels intended for high temperature service are the electric, such as those of the Heroult type, and the open-hearth. From results available on two plain carbon steels of essentially the same chemical composition and in the same heat treated condition, one of which had been made in the electric-arc furnace and the other in the open-hearth furnace, electric furnace steel would appear, as shown in Table VI, to possess the higher creep strength. This same condition has also been found to exist with a 0.50 per cent molybdenum steel at 1000 degrees Fahr. Too broad a conclusion is perhaps not justified from the limited data available but the work which has been done does point to the superiority of the electric-arc furnace product.

Creep values for steel made in the high-frequency induction furnace and in an electric-arc furnace are similarly compared in Table VII. In this case the induction steel appears to have creep

Table VII

Influence of Melting Process
Induction Versus Electric Furnace Steel

Material	Type Furnace	Temperature Degrees Fahr.	Stress for Designated Rate of Creep Per Cent per 1000 Hours		
			0.01	0.10	1.0
Mn-Mo-V	Electric	900	19,000	31,000	50,000
Mn-Mo-V	Induction	900	22,900	35,000	54,000
Mn-Mo-V	Electric	1000	5,500	11,800	24,000
Mn-Mo-V	Induction	1000	12,500	18,000	26,000

Note: Creep Tests of 1000 Hours Duration.

characteristics superior to those of the electric-arc furnace steel. It should be realized, however, that the induction-furnace steel was made in a furnace of 20 pounds capacity whereas the electric-arc furnace steel was made in a furnace of about 6 tons capacity. Since the differences in the respective sizes of these furnaces may likewise influence the creep characteristics, tests should be made in furnaces of comparable capacity before deductions are drawn.

Melting Practice—Melting practice also appears to have a considerable influence on the resulting creep characteristics. From the results given in Table VIII it appears that at temperatures of 900 and 1000 degrees Fahr. killed steel is decidedly superior to open steel. This condition is not pronounced at 800 degrees Fahr. and it is probable that it would not be found to exist to any greater extent at temperatures below 800 degrees Fahr.

There are two possible explanations of the differences in the creep characteristics of these two types of steel. One could be based on the assumption that the lowest temperature of recrystallization of

Table VIII

Influence of Melting Process
Creep Characteristics of Killed and Open Plain-Carbon Open-Hearth Steel at Temperatures of 800, 900 and 1000 Degrees Fahr.

Type	Temperature Degrees Fahr.	Stress for Designated Rate of Creep Per Cent per 1000 Hours		
		0.01	0.10	1.0
Killed	800	13,000	18,000	25,000
Open	800	15,000	19,500	25,000
Killed	900	11,250	15,250	20,500
Open	900	5,400	8,000	11,500
Killed	1000	3,200	6,800	14,500
Open	1000	1,800	3,550	7,000

Note: Creep Tests of 500 Hours Duration.

the open steel is lower than that of the killed steel. This assumption is substantiated by the fact that grain growth occurs at a more rapid rate in open than in killed steels.[9] Since creep is closely related to the lowest recrystallization temperature, the relative creep resistance of these two steels should depend upon the degree to which the given testing temperature is higher than this apparently critical temperature.

The second explanation assumes that because of the character of the open steel it contains a greater number of nonmetallic inclusions and gases, a considerable portion of which are located in or near the grain boundaries. Since creep is believed to be largely a phenomenon involving grain boundary slip at temperatures above the lowest temperature of recrystallization any weakening of the grain boundaries will result in a higher creep rate under a given load.

Casting Practice—It is also believed that the type and size of ingot into which the steel is cast may be found to have considerable bearing on the resultant creep characteristics. The size of ingot will determine the rate of cooling from the liquid to the solid state and this, in turn, will determine the size of the dendrites and the location of the metallic and nonmetallic constituents. It must also be kept in mind that another factor in addition to ingot size is involved. The second factor is the degree of reduction during hot fabrication and it is believed that this is just as important a variable as ingot size. The smaller the ingot the less will be the reduction during fabrication which, in turn, should influence the resulting creep characteristics. Wyman[7] found that size of ingot did influence the resulting properties with the smaller ingots tending to produce a more banded structure which, in turn, decreased the creep resistance. He further found that the influence of ingot size could be lessened if the smaller molds were preheated to certain temperatures.

Results obtained from our own laboratories indicate that ingot size is important. Specimens taken from a series of ingots varying in weight from 20 pounds to 3800 pounds have shown the stress required for a creep rate of 0.01 per cent per 1000 hours to vary from 8200 to 12,000 pounds per square inch with the smaller ingots showing the higher creep resistance. The differences, however, were not as marked for the higher creep rates. Sufficient information is not available to permit definite conclusions as to the exact influence

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[9]W. Oertel and A. Schepers, "Properties of Killed and Open Steels," *Stahl und Eisen*, June 1931, p. 710-715.

of ingot size but the results which have been obtained certainly do indicate that this is a factor which should be considered.

Grain Size—It is known that the size of crystals comprising the structure of the metal influences the creep characteristics. There are two aspects of the effect of grain size upon the behavior of steels. It is necessary to consider not only the actual grain size of the structure but also the "inherent grain size" which is established by means of a standardized carburizing treatment. What may be called the "structural grain size" refers to the size of the crystals that the structure happens to exhibit when the steel is examined. Inherent grain size is determined by heating samples of steel eight hours at 1700 degrees Fahr. in an active carburizing medium, cooling the samples slowly from the carburizing temperature, and counting the number of openings in the carbide network in one square inch of the case microstructure as viewed at 100 diameters magnification.

A numerical classification of steels for inherent grain size is presented in A.S.T.M. Standard A-19-33. By this system of rating the numbers 1 to 5 are identified as coarse-grained and the numbers 5 to 8 as fine-grained steels. Inherent grain size is a misnomer that cannot be displaced easily because the term is already in common use. The grain size numbers really measure the rate of grain growth at 1700 degrees Fahr. Each steel has only one rating for inherent grain-size but may present a large number of structural grain-size variations corresponding to the thermal history of the steel.

Inherent Grain Size—It has been found possible to correlate creep strength with the grain-size rating in the same manner as the other physical properties. In Figs. 4 and 5 the results of creep investigations are given, plotted to logarithmic coordinates of tests at 800, 1000 and 1200 degrees Fahr. on two different steels, each of which was tested in both the coarse-grained and fine-grained conditions. The coarse-grained material had an A.S.T.M. grain-size rating of 4-5 and the fine-grained a rating of 7-8.

With both of these analyses considered, the coarse-grained material possessed the greater creep resistance at 1000 and 1200 degrees Fahr. With the manganese-molybdenum steels the same condition existed at 800 degrees Fahr. indicating that, if the theories as to the nature of creep are correct, either the equicohesive temperature for this steel must lie below 800 degrees Fahr. or else marked differences in the strain-hardening characteristics may influence this relationship

at lower temperatures. With the molybdenum steel at 800 degrees Fahr. the fine-grained steel possessed the maximum creep resistance for the lower creep rate which substantiates the theory concerning creep. The equicohesive temperature of this carbon-molybdenum steel apparently lies between 800 and 1000 degrees Fahr.

Actual Grain Size—While many comments have been advanced concerning the influence of this factor on the creep resistance only a

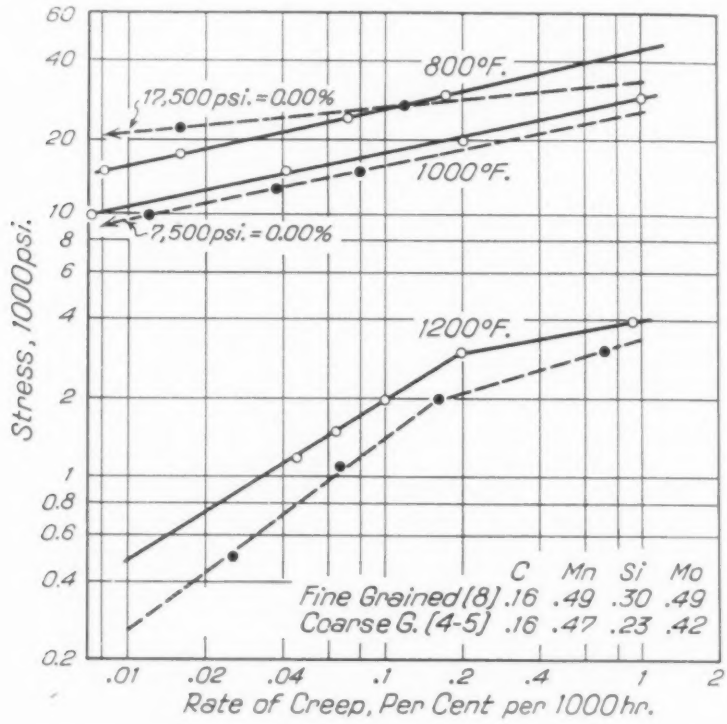


Fig. 4—Influence of Grain Size (Inherent) on the Creep Stress Curves of 0.50 Per Cent Molybdenum Steel at 800, 1000 and 1200 Degrees Fahr. Creep Tests of 500 Hours Duration.

few of them have been accompanied by substantiating data. On the basis of the theories advanced a coarse-grained material should possess the maximum creep strength at temperatures above the equicohesive temperature while at temperatures below the fine-grained material should be superior.

In 1932 the authors[10] presented data obtained from several nonferrous alloys, certain of which differed only in their grain size, at temperatures both above and below their lowest temperature of

[10] "Influence of Recrystallization Temperature and Grain Size on the Creep Characteristics of Nonferrous Alloys," *Proceedings*, American Society for Testing Materials, Vol. 32, 1932, p. 492-516.

recrystallization. The results, summarized in Fig. 6, do indicate that the influence of grain size varies, depending upon the temperatures being considered. At temperatures below the lowest temperature of recrystallization (300 degrees Fahr. for the 59-40-1 alloy and 400 degrees Fahr. for the 77-22-1 material) the fine-grained material

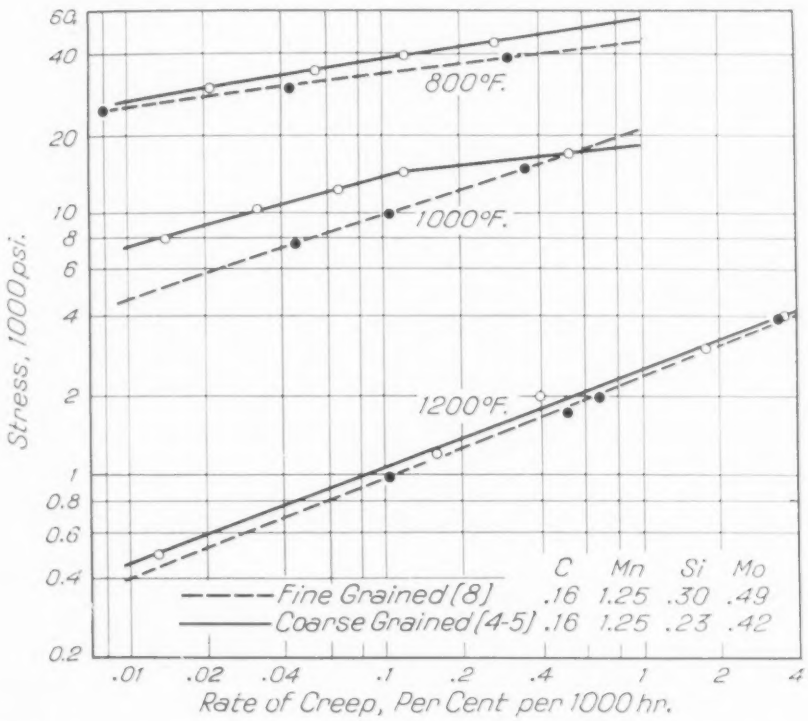


Fig. 5—Influence of Grain Size (Inherent) on the Creep Stress Curves of Manganese-Molybdenum Steel at 800, 1000 and 1200 Degrees Fahr. Creep Tests of 500 Hours Duration.

was superior; while at temperatures above (400 degrees Fahr. for the 59-40-1 alloy and 600 degrees Fahr. for the 77-22-1 material) the coarse-grained alloy had the greater creep strength.

Cross and Johnson[11] varied the grain size in a 4-6 Cr + Mo Steel by varying the heat treating temperature. They found that annealing from 2100 degrees Fahr. gave better creep resistance at 1100 degrees Fahr. than did annealing from 1550 degrees Fahr. Since the higher temperature treatment produced the larger grain structure it follows that in this case the coarse-grained steel possessed the greater creep resistance. While information was not given it is

[11] "Creep Properties of Oil Still Tubes," *Proceedings*, American Society for Testing Materials, Vol. 34, 1934, p. 80-100.

believed that 1100 degrees Fahr. would be above the lowest temperature of recrystallization for this material.

On the basis of these findings it appears that, from a creep strength standpoint, a coarse grain is to be preferred at the higher temperatures. Properties other than creep, however, must also be

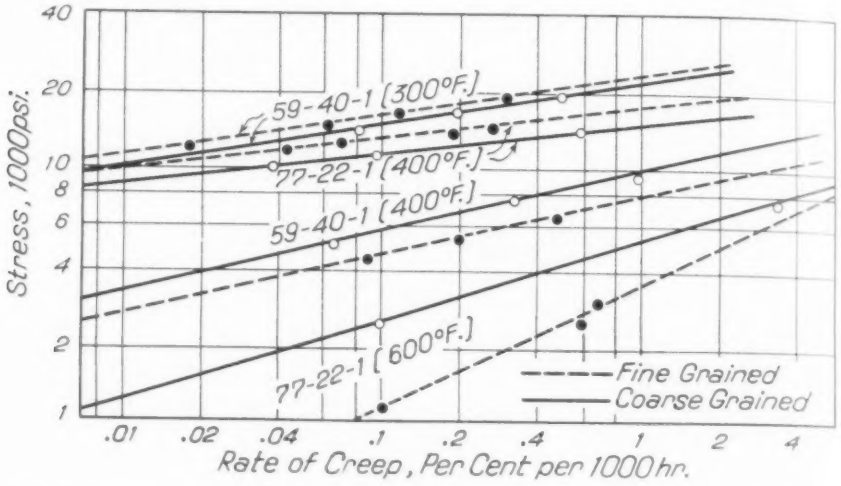


Fig. 6—Creep Stress Diagram for Copper-Zinc-Tin Alloys Showing the Influence of Grain Size on Rate of Creep. Creep Tests of 500 Hours Duration.

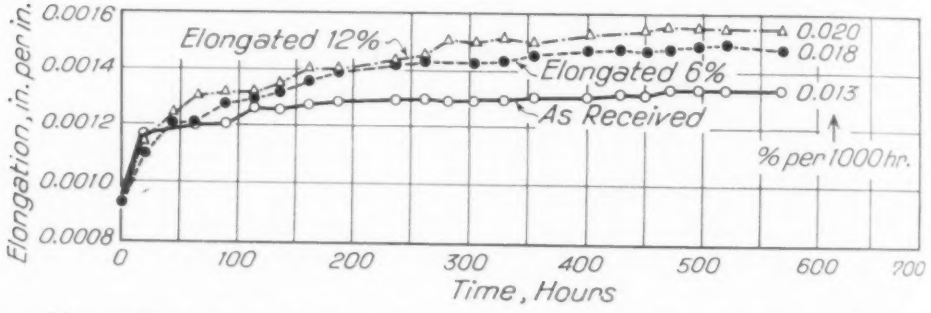


Fig. 7—Effect of Previous Deformation on the Creep Characteristics of Grade B Steel at 600 Degrees Fahr.

considered and oftentimes it may be necessary to restrict the grain size in order to obtain the best combination of properties.

Previous Deformation—This likewise is a subject upon which little information is available but the results which are indicate that it also influences the resulting creep strength. Results obtained from a 0.40 per cent carbon steel subjected to creep tests at 600, 850 and 1000 degrees Fahr. in the hot-rolled condition as well as after having been elongated cold 6 and 12 per cent clearly show the influence of this factor. These findings, shown in Figs. 7, 8 and 11, indicate that

at all three temperatures the hot-rolled material has the greatest creep resistance. The difference between the specimens was slight at 600 degrees Fahr. but became larger at 850 and 1000 degrees Fahr.

While hardness readings taken before and after the tests indicated no appreciable change, an X-ray examination showed that the hot-rolled material had undergone no visible change whereas the cold-deformed specimens did. The lattice changes after the

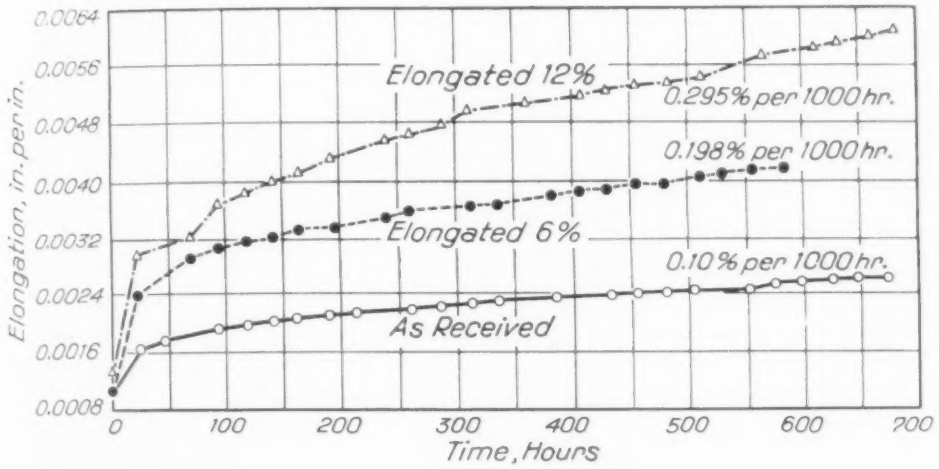


Fig. 8—Effect of Previous Deformation on the Creep Characteristics of Grade B Steel at 850 Degrees Fahr.

creep tests were slight at 600 degrees Fahr. in both the cold-worked specimens and at 850 degrees Fahr. in the specimen deformed 6 per cent. In the other cases the changes were pronounced.

The greater creep resistance of the unstrained material is, therefore, due to the fact that it possesses a more stable structure. An unstable structure implies that the crystals are in a state of greater mobility and will, therefore, deform more readily under an applied stress than if they were less mobile.

Equicohesive Temperature—If it is agreed that creep is a resultant of strain-hardening and recrystallization, it is obvious that the lowest temperature of recrystallization is of great importance. If the test temperature is below this critical temperature then the rate of strain-hardening will determine the creep characteristics; while if the test temperature is above, the relative rates of recrystallization and strain-hardening will predominate.

It is the authors' belief that the equicohesive temperature and

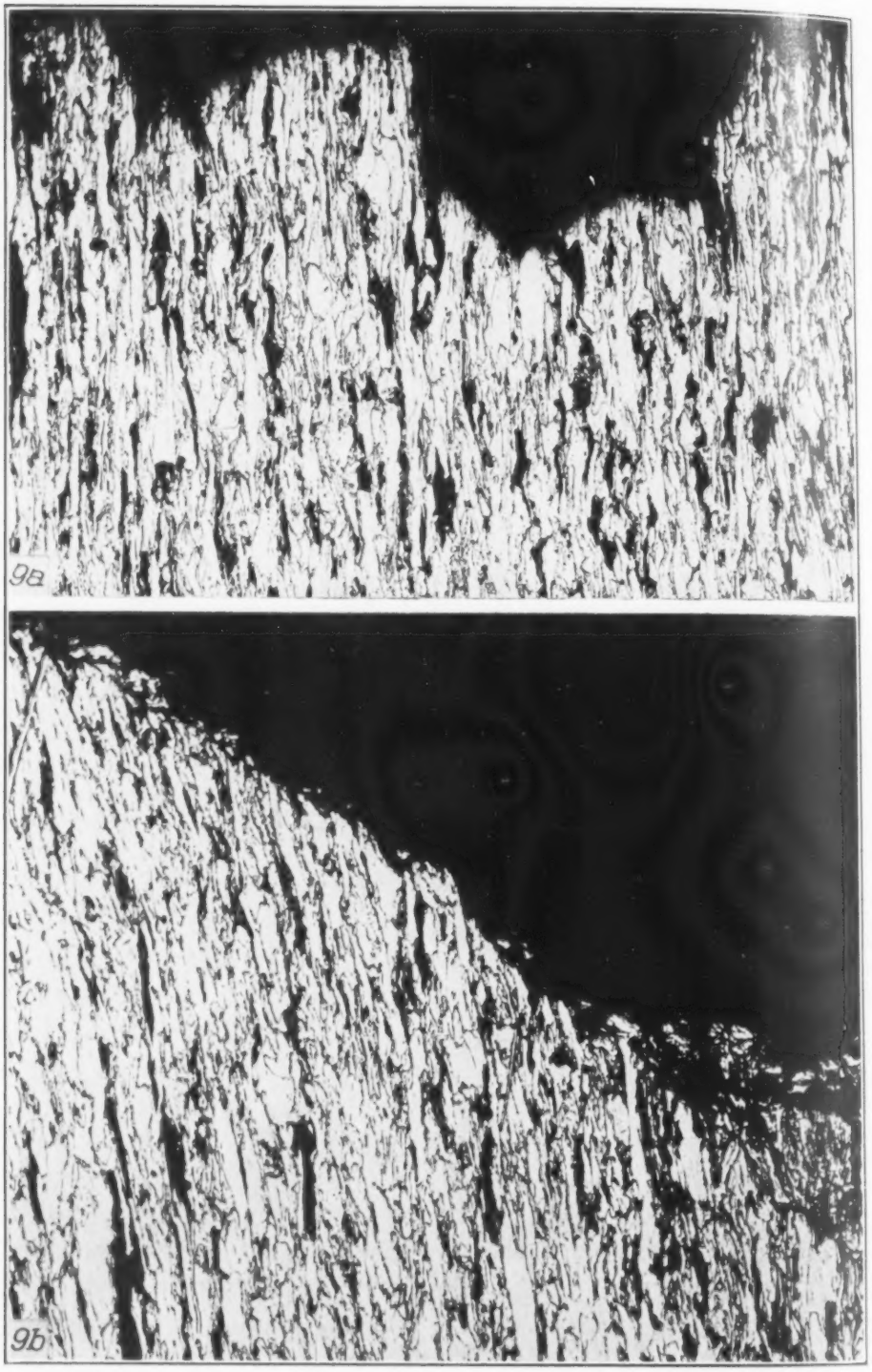


Fig. 9 a and b—Influence of Time and Stress on the Fracture of Electric Furnace S.A.E. 1015 Steel at 1000 Degrees Fahr. a—Short-Time Tensile Fracture. b—25,000 Pound Stress, 6.4 Hours for Rupture.

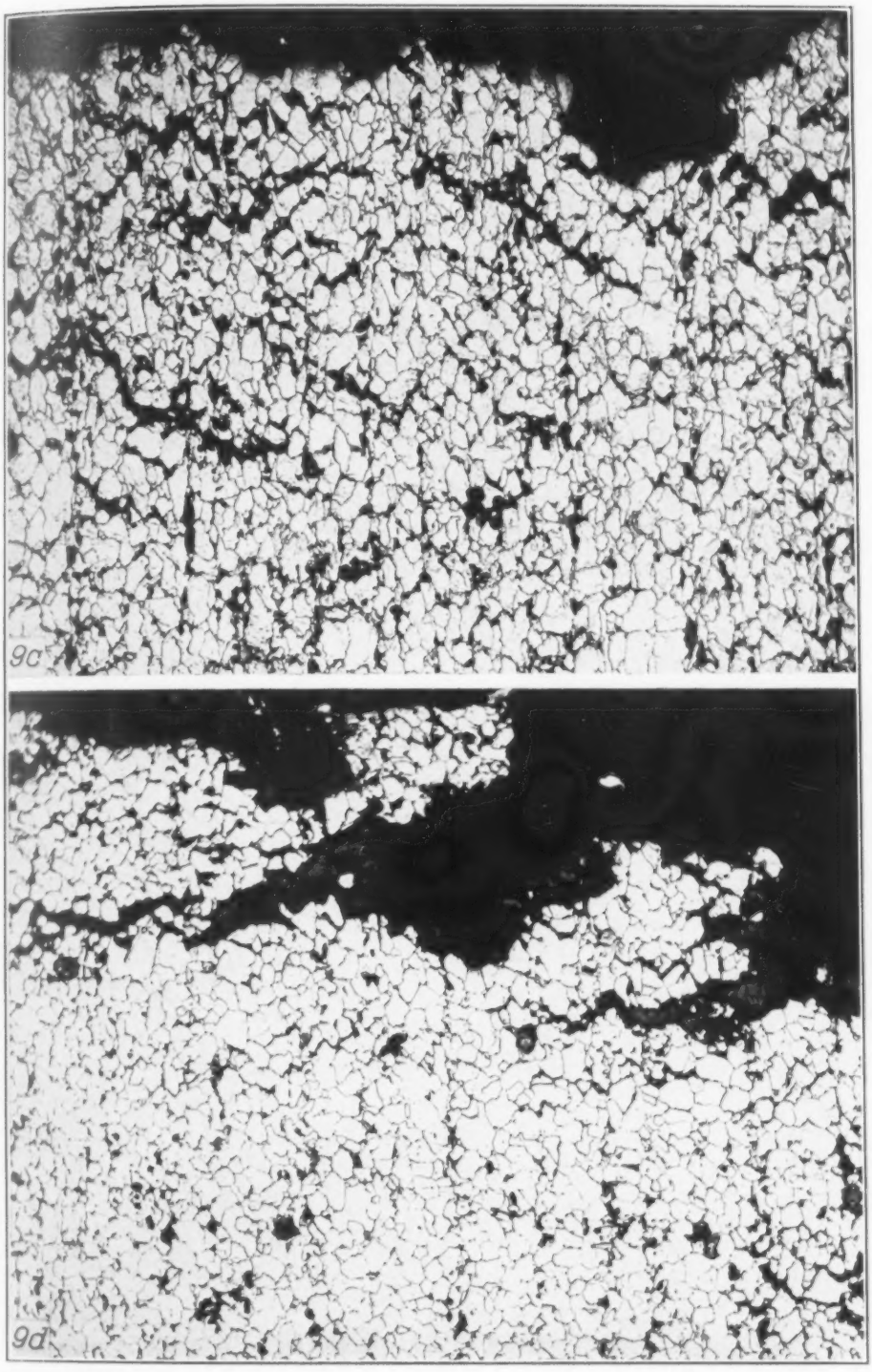


Fig. 9 c and d—Influence of Time and Stress on the Fracture of Electric Furnace S.A.E. 1015 Steel at 1000 Degrees Fahr. c—12,000 Pound Stress, 1552 Hours for Rupture. d—6000 Pound Stress, 13,950.75 Hours for Rupture.

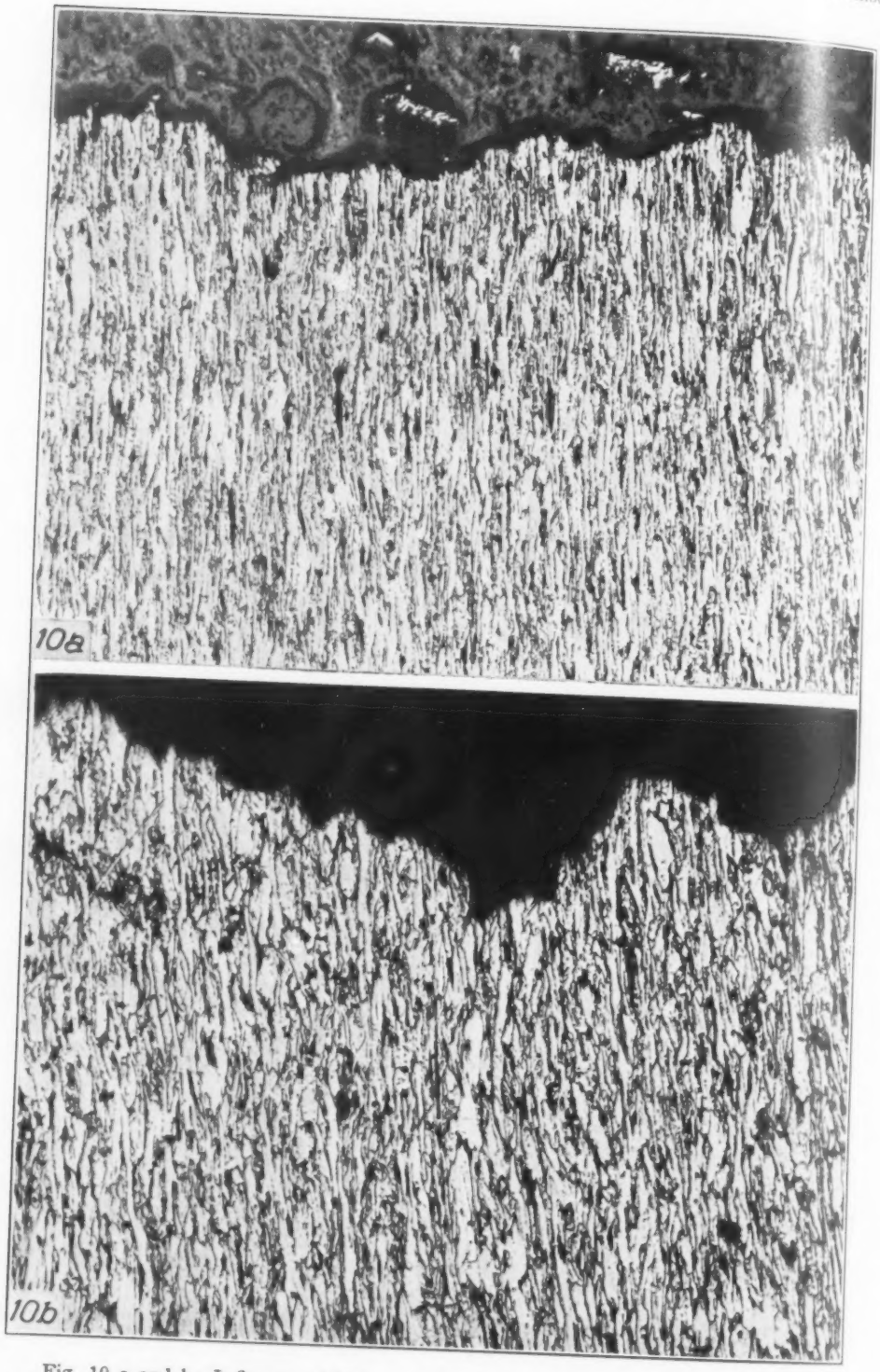


Fig. 10 a and b—Influence of Time and Stress on the Fracture of Cr-Si-Mo Steel at 1000 Degrees Fahr. a—Short-Time Tensile Fracture. × 100. b—34,100 Pound Stress, 2625.5 Hours for Rupture. × 100.

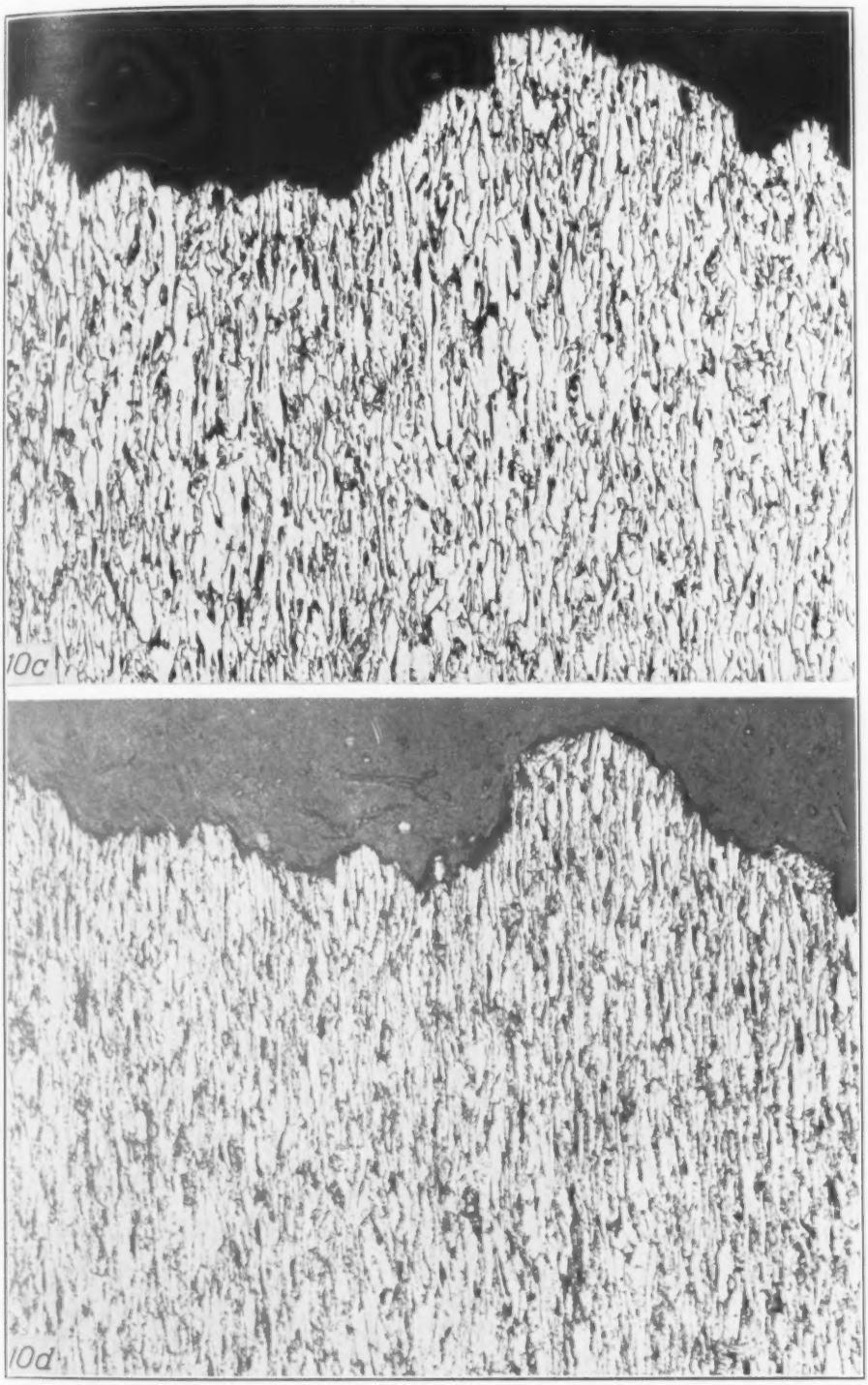


Fig. 10 c and d—Influence of Time and Stress on the Fracture of Cr-Si-Mo Steel at 1000 Degrees Fahr. c—30,000 Pound Stress, 3250 Hours for Rupture. × 100. d—24,600 Pound Stress, 6151 Hours for Rupture. × 100.

lowest temperature of recrystallization are one and the same. On the basis of the definition the equicohesive temperature is that temperature at which the grain boundaries possess sufficient mobility to allow deformation and thus fracture to occur through them. The lowest temperature of recrystallization on the other hand may be defined as the lowest possible temperature at which recrystallization may (not necessarily does) occur under the most favorable conditions. Both terms imply, therefore, that the grain boundaries possess sufficient mobility to allow the grains to reorientate themselves at the boundaries. Further, it is believed that this temperature is a constant for any given material.

If this assumption concerning creep is accepted then the differences in the creep characteristics of various materials with temperature can be readily explained. For example, lead whose lowest temperature of recrystallization is below room temperature exhibits the same general creep phenomenon at room temperature as plain carbon steels do at 900 to 1000 degrees Fahr. Likewise a 70-30 brass at 600 degrees Fahr. and a 60-40 brass at 400 degrees Fahr. are similar to carbon steel at 900 to 1000 degrees Fahr.

Table IX summarizes certain of the authors' previous results[10] obtained on nonferrous alloys. On the basis of these findings it appears that at temperatures below the lowest temperature of recrystallization the given materials can withstand appreciable stresses without measurable creep. Also, if the given structures are stable, the stress required for a creep rate of 0.01 per cent per 1000 hours at temperatures just below the equicohesive temperature corresponds rather closely to the proportional limit at this same temperature. Furthermore, at temperatures above the lowest temperature of recrystallization, creep occurs under very small stresses and it is believed that if sufficiently sensitive apparatus were available creep could be detected under any stress whatsoever.

Referring to Fig. 6 it also appears that the position of the test temperature with respect to the lowest temperature of recrystallization may change the slope of the curves produced when stress values are plotted against rates of creep on logarithmic coordinates. The slope undergoes a marked increase as the test temperature exceeds the lowest temperature of recrystallization.

While little information is available concerning the influence of alloying additions on the lowest temperature of recrystallization or equicohesive temperature of steels, it is believed that observed dif-

Table IX

Influence of Lowest Temperature of Recrystallization (Equicohesive Temperature)
Comparison of Proportional Limit and Creep Values for Nonferrous Alloys at Elevated Temperatures

Material	Temperature of Recrystallization	Temperature °F.	Temperature °C.	Proportional Limit Lbs./Sq.In.	No Measurable Flow	Stress for Designated Rate of Creep Per Cent per 1000 Hours 0.01	0.10	1.0
70-30	Over 400°F. (205°C.)	400	205	13,000	10,000	12,700	18,000	27,000
	Under 500°F. (260°C.)	600	315	6,750	a	290	850	2,150
		800	425	4,000	a	a	a	a
70-29-1	Over 400°F. (205°C.)	400	205	21,000	10,000	13,000	19,000	27,000
	Under 500°F. (260°C.)	600	315	15,000	a	1,000	1,950	3,800
		800	425	8,000	a	54	160	500
85-15	Over 500°F. (260°C.)	400	205	7,000	7,500	8,800	12,000	17,000
	Under 600°F. (305°C.)	600	315	5,500	a	1,000	2,600	6,800
60-40	Over 300°F. (150°C.)	300	150	7,500	7,500	9,000	12,000	17,000
	Under 400°F. (205°C.)	400	205	5,000	a	2,000	4,750	11,500
77-22-1	Over 500°F. (260°C.)	400	205	8,000	7,500	10,500	13,000	16,500
	Under 600°F. (315°C.)	600	315	4,000	a	1,200	2,500	5,300
59-40-1	Over 300°F. (150°C.)	300	150	14,000	10,000	12,000	15,000	21,500
	Under 400°F. (205°C.)	400	205	6,000	a	3,500	5,700	9,400

[a]Known to be very small and believed to approach zero.

Note: Creep Tests of 500 Hours Duration.

ferences in the creep characteristics of many alloy steels may be contributed to this factor. In fact, data available for certain alloys substantiate this belief. In Table I the alloy steel referred to as Cr-Si-Mo possesses, at 1000 degrees Fahr., a creep strength about five times that of the plain carbon steel. An examination of the fractures from extended-time creep specimens shows that for the carbon steel 1000 degrees Fahr. is above the equicohesive temperature, while for the alloy steel this same temperature is below the lowest temperature of recrystallization. The microstructures of certain of these fractures are shown in Figs. 9 and 10 and from these it will be observed that when sufficient time was allowed for fracture the structure of the carbon steel was entirely equiaxed and fracture occurred largely around the grain. With the alloy steel, however, regardless of the time required for fracture, the grains in the vicinity of the fracture were severely strained and the path of the fracture was not intercrystalline. The microstructures of these two steels support the contention that the lowest temperature of recrystallization and equicohesive temperature are the same.

Structural Stability—As discussed in connection with the influence of heat treatment on creeep there is a question as to whether or not any type of structure possesses complete stability at the more elevated temperatures. Lack of stability from a purely structural standpoint is generally manifested by carbide precipitation, spheroidization or structural transformation. Certain materials, moreover, whose structures undergo no visual change are also unstable as revealed by changes in their physical properties after service at elevated temperatures. An example of this latter condition is the phenomenon known as temper-embrittlement which may be defined as the loss of room temperature impact due to heating at certain elevated temperatures. While certain investigators claim that the cause of this embrittlement is due to sub-microscopic precipitation of carbides or nitrides, or to transformations in certain of the precipitated constituents, the question is still highly debatable.

It is further believed that the susceptibility of a material to intergranular oxidation or corrosion, even in the absence of visual structural changes, should be classified as a condition of structural instability. As an example, the case of a plain carbon steel at 1000 degrees Fahr. may be considered. If this material be subjected to stresses which will cause rupture over varying time periods it will be found, as shown in Fig. 12, that the ductility of the fractured

specimens decreases rapidly as the time required for rupture is extended. For instance in the case cited, the elongation of the short-time tensile specimen at 1000 degrees Fahr. was 42.5 per cent while that of the specimen which required approximately 14,000 hours for rupture was only 11.0 per cent. Microstructures of the surfaces of these specimens, certain of which are shown in Fig. 13, clearly

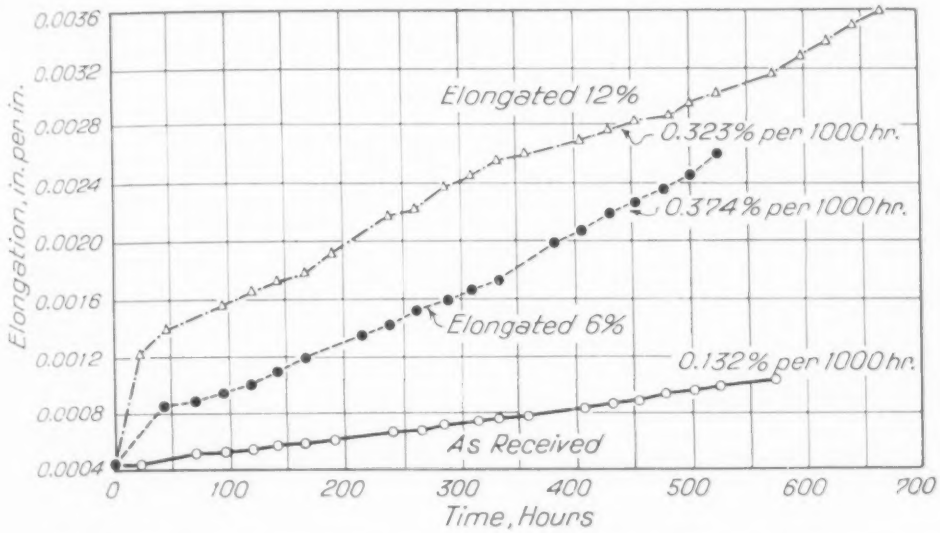


Fig. 11—Effect of Previous Deformation on the Creep Characteristics of Grade B Steel at 1000 Degrees Fahr.

show that this material suffered pronounced intergranular oxidation under the given test conditions and that this was the cause of the marked decrease in the ductility.

Pearlitic-type steels may, under certain operating conditions, undergo a structural change known as spheroidization which consists of the pearlite changing from a laminated to a spherical state. As shown by Bailey and Roberts[8] this change occurs in a relatively short time at temperatures of 1100 to 1300 degrees Fahr. and it has been assumed that the same change will occur under more prolonged times at the lower temperatures. In fact, these investigators have, on the basis of a logarithmic relationship, estimated the time required for temperatures as low as 750 degrees Fahr. The authors have produced spheroidization in a carbon steel at 1000 degrees Fahr. in the presence of stress, a factor which is known to hasten the phenomenon. As shown in Fig. 14, the original material contains pearlite which was largely although not completely laminated. Approximately 5000 hours at 1000 degrees Fahr. under a stress of 9000 pounds caused

appreciable spheroidization while 14,000 hours at this same temperature under a stress of 6000 pounds not only completely spheroidized the pearlite but also produced considerable migration of the spheroidized carbides.

The influence of spheroidization on the physical properties is not completely established. It is believed that in the majority of cases the creep resistance is reduced although, as shown in Table IV, certain exceptions exist. It also tends to lower the tensile strength and to increase the ductility and impact resistance.

If during service the steel has been heated above its critical range structural transformation will occur. The resulting phase changes may affect the grain size of the structure, the character of

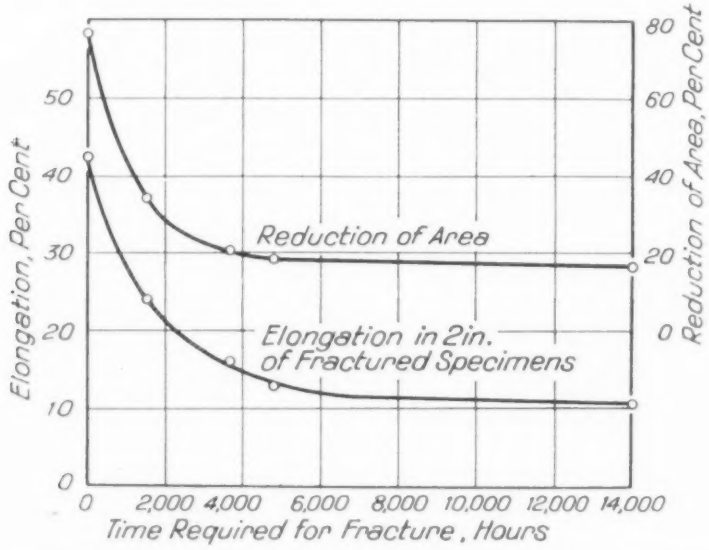


Fig. 12—Structural Instability. Influence of Time Required for Rupture at 1000 Degrees Fahr. on the Ductility of 1015 Carbon Steel.

the constituents, and the mechanical properties of the steel while hot, or subsequently when it has been cooled to atmospheric temperature.

Certain of the highly alloyed ferritic type steels lack structural stability in that at the higher operating temperatures they may undergo marked grain growth with a resulting loss of ductility and impact resistance. Since many of these steels do not possess a definite critical range the normal structures cannot be restored by the usual heat treatments.

Lack of structural stability in many of the austenitic type steels is manifested through carbide precipitation and migration of the precipitated carbides to the grain boundaries. This precipitation at

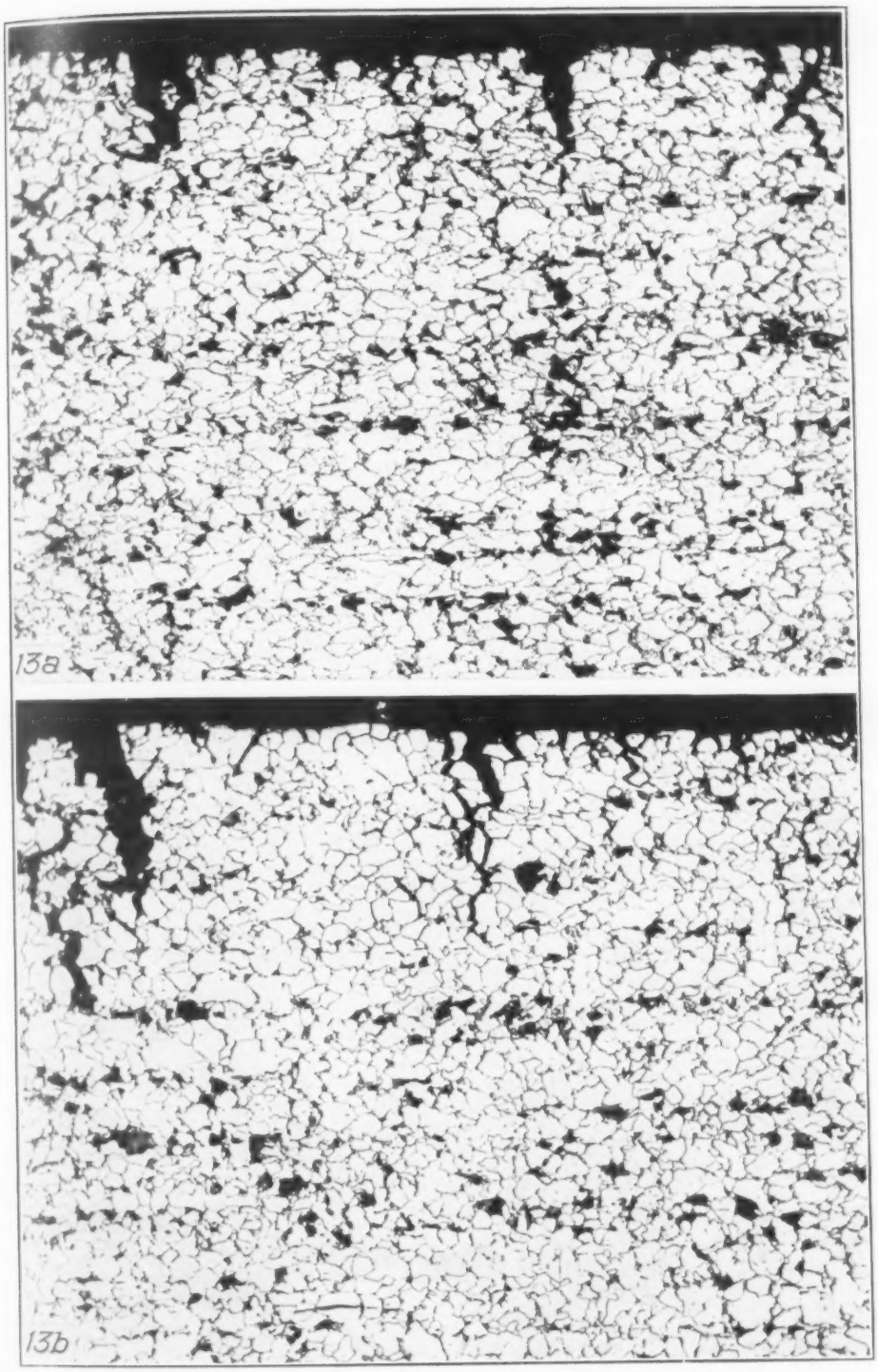
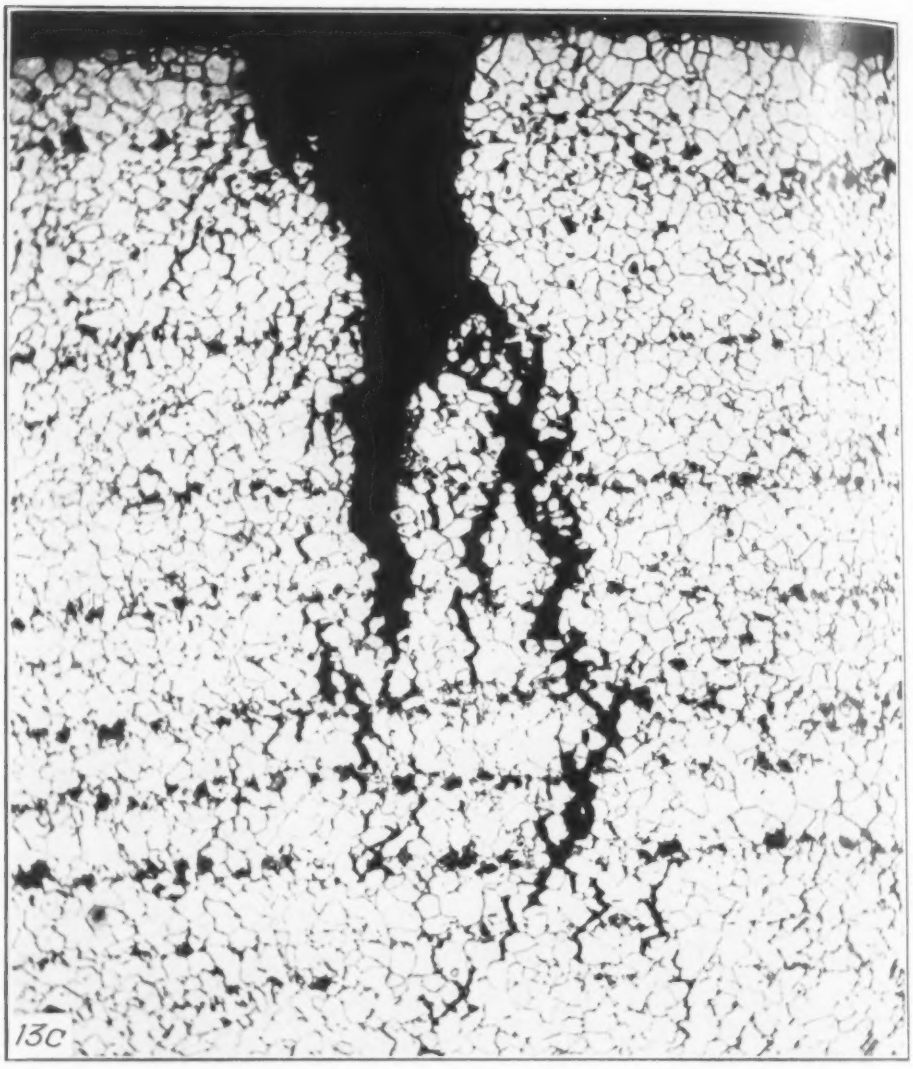


Fig. 13—Structural Instability. Microstructure Showing Intergranular Oxidation of 1015 Carbon Steel at 1000 Degrees Fahr. × 100. (a) 12,000 Pound Stress, 1552 Hours for Rupture. (b) 9000 Pound Stress, 4788.5 Hours for Rupture. (c) 6000 Pound Stress, 13,950.75 Hours for Rupture. (See Next Page for Fig. 13c.)



See Page 863 for Description of this Photomicrograph.

the grain boundaries is considered to promote intergranular corrosion and oxidation and may likewise cause appreciable loss of ductility and impact resistance. Its influence on the creep strength is not definitely known although Norton[7] has reported it to be without appreciable influence for the particular case he investigated. Even though the actual creep strength may not be influenced, information available indicates that the total amount of deformation the material may undergo before fracture is decreased by this phenomenon and actual fracture may, therefore, occur without pronounced deformation and thus no indication of impending failure will be given.

Fig. 14—Structural Instability. Spheroidization of 1015 Carbon Steel at 1000 Degrees Fahr. (a) Original Annealed Structure. × 1000. (b) Creep Specimen at 1000 Degrees Fahr., 9000 Pound Stress, 4788.5 Hours. ×1000. (c) Creep Specimen at 1000 Degrees Fahr., 6000 Pound Stress, 13,950.75 Hours. × 1000.

Conclusions

On the basis of the results presented it must be concluded that the ability of a metal to resist creep is a sensitive property which is greatly influenced by many factors that have little or no influence on the remaining physical properties usually considered.

In other words, there are many factors other than composition which affect the creep resistance and check results will not in general be obtained from steels in which only this factor had been maintained constant. Before values obtained from one steel may be applied to another of the same composition it is necessary to know the complete past history of both steels.

The influence of the various factors which were considered may be briefly summarized as follows:

Chemical Composition—The desired operating temperature determines the most suitable composition. At temperatures below the lowest temperature of recrystallization (equicohesive temperature) the creep resistance may be increased either by certain elements which largely enter into solid solution in the ferrite, such as nickel and manganese, or by the carbide-forming elements, such as chromium, molybdenum, tungsten and vanadium. At temperatures above the lowest temperature of recrystallization, however, the carbide-forming elements are the most effective in increasing the creep strength. In fact, at these more elevated temperatures the creep resistance of pearlitic type steels is often decreased by the addition of the non-carbide-forming elements.

Structural Uniformity—Nonuniform distribution of the various constituents will in general decrease the creep resistance, since the strength of the steel as a whole will depend upon that of the weakest portion of the mass.

Heat Treatment—The heat treatment should be such that a high degree of structural stability results at the given operating temperatures. For the higher operating temperatures annealing is in general the most satisfactory although a normalizing or quenching treatment can be used provided it is followed by a drawing temperature approximately 200 degrees Fahr. above the operating temperature. A normalized and drawn structure possesses in general a greater creep resistance than an annealed one, although annealing generally produces greater creep resistance, at the more elevated temperatures, than oil or water quenching.

Method of Manufacture—The creep resistance is likewise influenced by the melting process, the melting practice and the casting practice. Results available indicate the electric furnace steel to be superior to the open-hearth product, while the induction furnace material is superior to the electric furnace steel. For any given melting process, killed steels are superior to those of the open or rimmed type.

Grain Size—The influence of both actual and "inherent" grain size on the creep resistance varies depending on the temperature being considered. At temperatures below the lowest temperature of recrystallization, a fine-grained steel (either actual or inherent) possesses the greater creep resistance, while at temperatures above, a coarse-grained structure is superior.

Previous Deformation—Initial cold deformation decreases the creep resistance at temperatures above the lowest temperature of recrystallization with the differences being more pronounced at the more elevated temperatures. Under sufficiently prolonged testing periods, however, these differences would probably be largely eliminated.

Equicohesive Temperature—Equicohesive temperature, or lowest temperature of recrystallization, is believed to be critical insofar as the creep resistance is concerned. At temperatures below this, strain-hardening predominates and stresses of appreciable magnitude can be withstood without continuous creep. At temperatures above the equicohesive temperature, however, the rate of recrystallization will exceed the strain-hardening rate and continuous creep will occur under very low stresses.

Structural Stability—In general it may be stated that the greater the structural stability, the greater will be the creep resistance. As pointed out, however, there is a question as to whether any type of structure possesses complete stability especially at the more elevated temperatures.

In conclusion it should be emphasized that while this paper has been largely confined to a discussion of the creep characteristics, this property alone does not necessarily determine the complete suitability of a material for high temperature applications. Proper consideration must also be given to the remaining physical properties, such as oxidation and corrosion resistance, ductility, and impact strength if entirely satisfactory results are to be obtained with these materials.

ACKNOWLEDGMENTS

The authors desire to express their appreciation to The Timken Steel and Tube Company for the financial assistance which has made these investigations possible and for the permission to publish the findings. Particularly do they wish to express their gratitude to F. J. Griffiths, H. H. Timken, Jr., and R. L. Wilson. Also they wish to express their appreciation to J. W. Freeman for the metallographic studies.

DISCUSSION

Written Discussion: By F. H. Norton, Massachusetts Institute of Technology, Cambridge, Mass.

The authors of this paper are to be congratulated in assembling such a large amount of creep data on various types of steel in such a way as to bring out the importance of variations in the metal structure which cannot be detected by our ordinary examinations. All experimenters in this field have noticed the considerable variation in creep results between various heats of the same steel, which show identical structure and physical properties in our ordinary tests. Many of the discrepancies which in the past have been laid to lack of precision in testing methods are now believed to be actual variations in the metal itself, which as yet cannot be detected in any other way. The authors bring out the very real differences in creep properties of steel manufactured by different methods.

The data in this paper allow us to judge the value of various alloying constituents in steel as influencing the creep. The picture presented would lead us to believe that there are still great opportunities to produce alloy steels of even better creep resistance than we have at present.

It is undoubtedly true that the treatment which produces in the quickest time a final stable structure will be the best. There is some doubt in my mind, however, if such a final condition can be arrived at without taking a considerable length of time. We have found in our creep work that specimens annealed for 1000 hours, at, or slightly above the test temperature, reach a uniform condition of flow more rapidly than specimens annealed a short time.

Any criticisms that might be made of this paper would perhaps be directed at the fact that a considerable portion of the results have been obtained with creep tests lasting only 500 hours. I feel that such a length of time is insufficient to obtain any reliable data at the low rates of creep, and I am more and more coming to the conclusion that if we are to measure rates of the order of 0.01 per cent in 1000 hours, tests of at least 2000 hours duration are necessary. For example, referring to Fig. 7 in the paper, the curves undoubtedly give the relative total elongations with fair precision, but it is difficult to see how any definite creep rate can be determined from them. A considerable proportion of our tests are being carried to more than 5000 hours. This is necessary for those steels showing delayed structural changes which can materially increase the flow rate.

Oral Discussion

E. S. DAVENPORT:[1] I would like to ask Dr. Clark a question. I noted that he used the terms, "inherent grain size" and "actual grain size" during his presentation, and I wonder if he would be willing to elaborate a little bit for us on what is meant by these two terms,—what is the difference between inherent grain size and actual grain size?

Authors' Reply

C. L. CLARK: We wish to thank Professor Norton for his comments to our paper and are pleased to note that in general he is in agreement with the data presented.

Professor Norton raises one interesting point concerning the length of time creep tests should be conducted. As pointed out certain of the results presented were obtained from tests of only 500 hours duration. These particular tests date back for several years as all of our recent ones have been conducted for a minimum time of 1000 hours. The question now arises, however, whether shorter time tests, say up to 500 hours, are not sufficient for detecting differences of the magnitude indicated in this paper. As Mr. Norton also indicates, considerable variations may exist in different heats of the same analysis. If this is true, and all available information indicates that it is, are we justified in going to very extended testing time periods unless it is certain that the material eventually to go into service is not only of the same composition as the creep specimen but also from the same heat. This problem becomes even more complicated when it is realized that differences in creep resistance may also exist in the same heat. It may be that for most practical problems the degree of refinement has been carried to extremes. In other words, it may be similar to using a microscope where a one inch rule might be more adaptable.

With reference to Mr. Davenport's question "inherent" grain size is applied to that grain size resulting after the standard McQuaid-Ehn carburizing test. Actual grain size, on the other hand, is that resulting after the actual heat treatment to which the material was subjected.

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[1] U. S. Steel Corp. Research Laboratories, Kearny, N. J.

INTERPRETATION AND USE OF CREEP RESULTS

By J. J. Kanter

Abstract

The acceptance of creep as a factor in the design of structure for operation at high temperature necessitated a critical interpretation of laboratory test results. The character of the constant load creep strain-time curve is considered. Attention is directed to the various factors believed to influence creep rates. Evidence that creep obeys a viscous flow law at small stresses is considered. Data appear to indicate that creep viscosity decreases exponentially with increase of stress and the constant creep rate varies with stress accordingly. Bailey's work on creep by shear is reviewed, with suggestions upon the interpretation of creep under combined stresses. The subject of relaxation due to creep introduces problems in the choice of bolting materials. Selection of materials for high temperature application should be made upon creep tests showing position on the scale of temperature with regard to working stress. Working stresses must be determined upon a strict interpretation of creep tests with regard to total deformation tolerable.

DURING the past decade much attention has been directed upon creep of metals as a factor in the design of equipment for high temperature service. Prior to this period engineers were just beginning to appreciate the importance of creep through the pioneer investigations of Chevenard, Dickenson (1)[1] and French. At that time there was considerable reluctance on the part of many designers to admit the creep factor into consideration, due no doubt to their total lack of methods, practice and experience in such direction. It had been demonstrated that steels are of an intrinsically different mechanical character at even mildly elevated temperatures than at atmospheric. At ordinary temperatures, steel is an excellently elastic body. So long as structures sustain only elastic deformation they are of reasonably sound design. But it is now generally recog-

[1]The figures appearing in parentheses pertain to the references appended to this paper.

A paper presented as part of the Symposium on the Plastic Working of Metals, Eighteenth Annual Convention of the Society, Cleveland, October 19 to 23, 1936. The author, J. J. Kanter, is associated with Crane Co., Chicago. Manuscript received June 18, 1936.

nized that at elevated temperatures no definite limit of elasticity is determinable, and any practical limit which could be chosen upon this basis would be too small for designing equipment in an economic way upon a theory of stress-strain equilibrium. The early conception of a limiting creep stress, still entertained by some metallurgists, was perhaps fathered by the desire to avoid the difficulty of abandoning the established rules for choosing working stresses. But limiting creep stresses of magnitudes practical to design have not proven acceptable. Engineers have therefore become reconciled to the fact that creep must be accepted and dealt with if reasonable metal sections are to be employed at high temperatures.

Some applications permit more deformation than others, thus involving due consideration of the length of service which the equipment is expected to render. Corrosion, oxidation and other deterioration, as well as the time deformation factor enter into the dimensioning of load carrying members and fixing their stress. Such considerations have led to some serious attempts to develop rational bases for dealing with creep phenomena in design.

There naturally has been considerable reluctance among conservative engineers to use higher temperatures and pressures, attributable not only to lack of data and information concerning the long continued performance of alloys under extreme conditions, but, in addition, the lack of practical theory pertaining to the problems involved. This policy is due in a large measure to a feeling that a considerable risk is involved in attempting practical applications of laboratory findings.

The interpretation of long-time tensile or creep data has been and still is a matter of some controversy. In the early days of creep testing much effort was expended in attempts to justify limiting creep stresses in the classical sense, i.e., a stress below which no creep takes place at elevated temperature. As already intimated, it is now the generally accepted point of view that at elevated temperatures creep is present in some degree, at any practical stress—something which must be tolerated up to limits fixed by operating and other service conditions.

That the creep factor is recognized as of prime importance in high temperature design is witnessed by the already sizable literature pertaining to creep application. The list of references appended to this paper forms a bibliography of numerous publications devoted to the mathematical and engineering theory as well as experiments

developed for the treatment and application of creep data. The purpose of this paper is to present a summary of some of the many considerations bearing on the utilization of creep data. In preparing this digest, however, it has been necessary to neglect much important material represented in the bibliography.

The Constant Load Creep Strain-Time Curve

Basic tensile creep information as it is usually observed is represented by the constant load and temperature creep strain-time curve, of which there are two types, which may be classified

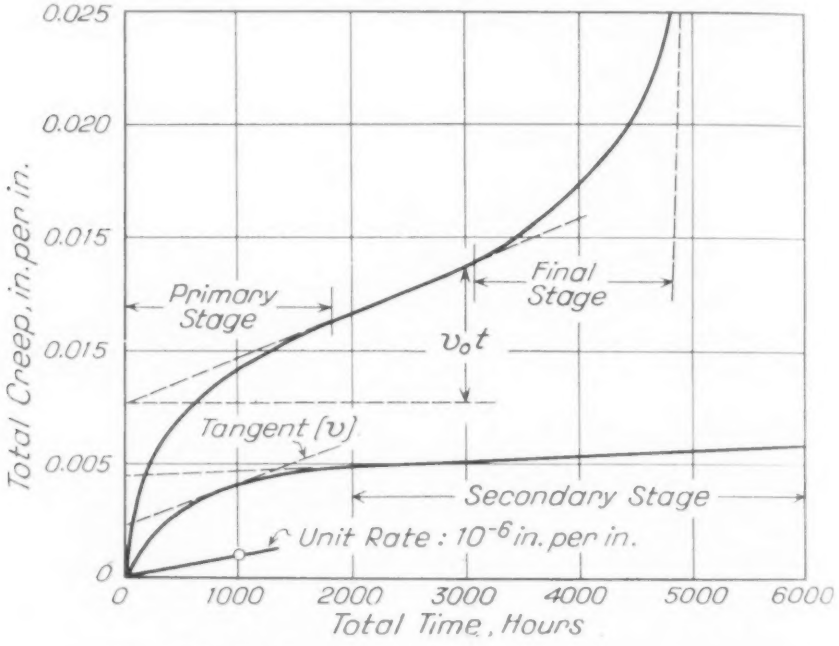

Fig. 1—Typical Creep Curves, Illustrating Primary, Secondary and Final Stages of Creep.

upon a temperature basis. At sufficiently high temperatures, where deformation is not associated with strain hardening effects, the usual creep curve follows an essentially straight line during a substantial testing period. In the lower ranges of temperature where creep effects are recorded, a continuously decreasing creep rate is shown throughout the test. This tendency to decrease is associated with strain hardening. The minimum temperature at which a constant rate of creep is found should be then the temperature at which strain-hardening relief proceeds at the same rate as it is incurred.

In Fig. 1, illustrating typical creep curves, it is seen that an initially higher rate of creep is associated with the curve for constant creep rate. This effect has been attributed, by Bailey and Roberts (2), to uneven stress distribution among the metal grains. Upon such a conception, the grains composing the bar of metal vary as to the stress they carry and consequently deform and strain harden at varying rates. The grains most highly stressed thereby tend to creep more rapidly than others. The stress relief effected by such deformation throws an excessive share of load upon neighboring grains. The process is thereby repeated with a tendency toward a more uniform distribution of stress. Such reorientation is accompanied by a falling off of the rate of creep and seems to be the predominating change which occurs during the first phase of creep. The second phase, characterized by a uniform rate of creep, attains, therefore, under the establishment of uniform conditions throughout the material. During this second phase of creep the area of section remains sensibly constant. Beyond certain limits of deformation, however, resistance to creep diminishes, leading to an increase of creep rate. This third phase of creep is characterized by accelerating creep rate, stress intensification, and necking of the section, leading to fracture.

Equations for the Time-Creep Curve

Much evaluation of creep curves has been made upon the assumption that the second phase of creep proceeds at a constant rate for periods many times greater than the actual periods of observation. As a first approximation and an expedient, this practice has proven helpful but, if a more precise equation for the creep-time curve is sought, additional considerations and refinements are necessary. Among the serious attempts that have been made in this direction is McVetty's (3) treatment upon the basis that the creep velocity in excess of a minimum rate decreases very nearly geometrically in accordance with an exponential relationship,

$$v - v_o = ce^{-at} \qquad [1]$$

where v_o is a constant and $v = \dfrac{d\epsilon_p}{dt}$ the slope of the creep curve.

The unit creep strain, ϵ_p, by integration becomes

$$\epsilon_p = \epsilon_o + v_o t - \frac{c}{a} e^{-at} \quad [2]$$

for time, t. a, c, and ϵ_o are constants.

This expression is equivalent to stating that a constant creep rate is approached asymptotically. McVetty shows that this function, with properly determined constants, represents actual creep curves with a good degree of approximation. For extrapolations or predictions of deformations in times greater than actually tested, he suggests the asymptote of equation [2] as a basis

$$\epsilon_p = \epsilon_o + v_o t \quad [3]$$

the constant ϵ_o representing the intercept on the strain ordinate. Inasmuch as this scheme of evaluation is invalid for deformations extending beyond the inflection toward the third phase of creep, McVetty recommends that extrapolations based upon it be carried no farther than the equivalent of one per cent elongation.

Creep Recovery

Removal of stress from a metal which has undergone creep strain has been shown to result in some creep recovery. This recovery, as illustrated by Fig. 2, proceeds at a diminishing rate approaching zero. Tensile observations upon this effect in iron and nickel

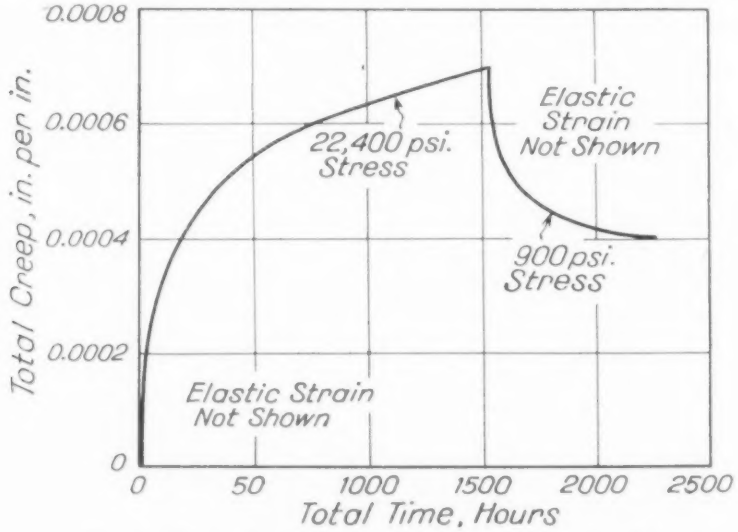

Fig. 2—Creep Recovery, Tapsell and Prosser (5) Data for Nickel-Chromium-Molybdenum Steel at 450 Degrees Cent.

wires have been reported by Chevenard (4). Tapsell and Prosser (5) measured recovery in a nickel-chrome-molybdenum steel at 840 degrees Fahr. (450 degrees Cent.) over a period of 1000 hours subsequent to stressing at 22400 pounds per square inch for 1500 hours. The creep strain of 0.0007 inch per inch reduced to 0.0004 inch per inch with the stress reduced to 900 pounds per square inch. Bailey (6) observed a considerable creep recovery in carbon steel tubing after deformation by combined torsion and tension stressing.

Whether creep recovery is a factor to be considered in connection with expressions for the creep strain-time function depends upon the interpretation of the effect accepted. McVetty (7) is of the opinion that careful study is required to distinguish between actual creep recovery and changes due to lack of stability of the metal itself.

It is well known to investigators of the creep of steels that when a creep bar is cooled and unloaded subsequent application of the same load and temperature returns the steel to precisely the same stage of creep previously attained, that is, no break in the creep curve results. Evidently creep recovery will not take place in cold steel, or, upon restressing, it would resume creep at a lower stage. Thus it seems reasonable to suppose that creep recovery is associated in some degree with strain-hardening relief. As such it might be regarded as a reversible component of the first phase of creep. Chevenard (4), however, considers the first phase creep strain-time curve as the sum of two components of deformation which he characterizes as permanent viscous and sub-permanent, the latter being associated with the creep recovery effect.

The Declining Creep Rate

A number of creep strain-time curves for steels at temperatures from 750 to 1000 degrees Fahr. (400-540 degrees Cent.) show a continuously decreasing creep rate extending over periods of a year or longer, suggesting complete cessation of creep eventually. Some of this data when plotted with strain and time ordinates both logarithmic is strikingly linear and may be represented by an equation in which creep strain is a power function of time,

$$\epsilon_p = at^n. \qquad [4]$$

$$\frac{d\epsilon_p}{dt} = v = ant^{(n-1)}. \qquad [5]$$

The exponent n of these relationships furnishes an index to strain-hardening under creep. When the value of n is unity (unit slope on log-log plot) the constant v_o is then the rate of uniform creep and obviously strain-hardening is not an influence. Values of n approaching zero can be interpreted as rapid cessation of creep due to strain-hardening. Data revealing increasing creep rates give n a value greater than unity. Some evaluations of n by Kanter and Spring (8) for steel of various compositions and treatments showed that at a constant temperature its value fluctuates slightly over a range of stress, but varies directly as temperature, ferritic steels showing values of 0.0 to 0.2 at 550 degrees Fahr. increasing to 0.8 to 1.0 at 1200 degrees Fahr. (650 degrees Cent.).

Plotting creep strain-time data using a linear strain ordinate and a logarithmic time ordinate is another method useful for analyzing strain-hardening influence upon creep, giving a straight line correlation of measurements in certain tests over lengthy test periods as shown by Tapsell and Prosser (5). Insofar as creep is represented by such a line, we have the approximate rule that the rate of creep is inversely proportional to the time of constant stress application, since

$$\epsilon_p = a \log (1 + bt). \qquad [6]$$

$$\frac{d\epsilon_p}{dt} = v = \frac{ab}{1 + bt}. \qquad [7]$$

Chevenard (4) chooses to represent both the first and second phases of creep in a single expression simply by adding a term dependent upon the constant creep rate to equation [6],

$$\epsilon_p = a \log (1 + bt) + v_o t. \qquad [8]$$

$$\frac{d\epsilon_p}{dt} = v = \frac{ab}{1 + bt} + v_o \qquad [9]$$

The number 1 in equation [9] has a negligible effect upon the value of v except for small values of t and may be dropped for computational purposes, giving a simpler expression,

$$\frac{d\epsilon_p}{dt} = v = \frac{a}{t} + v_o \qquad [10]$$

Chevenard (4) approximates the value of minimum creep rate v_o

by plotting $\frac{\triangle \epsilon_p}{\triangle t}$ versus reciprocal time which he finds to be a linear relationship. The value of v_o is obtained by extrapolating this line to intersect the zero $\frac{\triangle \epsilon}{\triangle t}$ ordinate, corresponding to $t = \infty$. This procedure is of interest in connection with the attempts that have been made to evaluate creep limits from short duration to creep tests.

The Constant Creep Rate

As suggested by the Bailey and Roberts (2) conception of differentiation between the first and second phases of tensile creep, a constant minimum creep rate attains as a result of a balance between a declining creep rate and an increasing creep rate. The declining creep rate is associated with stress redistribution among the grains and strain-hardening due to grain fragmentation. The increasing creep rate is associated with strain-hardening relief, stress intensification due to decrease of section, and other effects contributing to thermal instability. The idea has been advanced by Foley (9) and others that due to these counterbalancing effects creep is a cyclical process, the rate of creep undergoing slight fluctuations in accordance with the net influence of the various factors at work. While this idea is quite tenable, experimental technique has not as yet been refined to a point where the cyclical fluctuations of creep rate can be convincingly demonstrated.

Although the McVetty and Chevenard devices of representing the creep curve as the integral of the sum of constant and declining creep rates can have only an empirical significance in view of the complex picture of creep mechanism, there is a rational physical basis for such functions.

McVetty (3) points out that the assumption of a straight line asymptote is equivalent to an assumption of viscous flow in the second phase of creep and that if the viscosity coefficient for pure shear for a constant stress remains constant, the creep rate should approach a minimum value which will also remain practically constant over a considerable period. The experimental justification for this view, it would seem, depends upon establishing that creep takes place without change of density in the metal and the consequent potential energy changes affecting viscosity. Now, it has been demonstrated through

creep tests of single crystals of metal that the deformation proceeds largely by means of slip along the crystal planes subjected to maximum shear stresses. Lead crystals investigated by Betty (10) deformed most rapidly when the octahedral planes coincided with maximum shear planes. Similarly zinc crystals investigated by Miller (11) having their basal planes oriented with maximum shear planes, showed lowest creep resistance. Moreover single crystals so oriented have an exceedingly small elastic range. Bailey's (12) analyses of creep strains in tubes have demonstrated that polycrystalline metals also deform by shear. Creep by shear, with factors such as strain-hardening and stress readjustment remaining constant, should therefore be of an essentially viscous nature permitting the creep rate to remain practically constant at the minimum value for an appreciable length of time.

The Accelerating Creep Rate

How long tensile creep might continue at the minimum rate for constant load and temperature, granting equilibria of the stress distribution and strain-hardening factors, raises the question of the effect of stress intensification due to the decreasing cross-sectional area of the bar. Assuming no density change and approximately uniform reduction of area along the bar, the average stress intensity during creep extension will be

$$\sigma = \sigma_o (1 + \epsilon_p), \qquad [11]$$

where σ_o is the original stress intensity. Now, the relationship between minimum creep rate and stress at constant temperature can be accurately expressed for correction purposes over a limited range by the power function.

$$v = a\sigma^n. \qquad [12]$$

It follows then from [12] that the minimum rate of creep v should increase according to

$$v = v_o (1 + \epsilon_p)^n. \qquad [13]$$

Equation [13] serves then to show the magnitude of the effect which change of section alone might be expected to have upon increasing the minimum creep rate. In Fig. 3 the per cent increase of creep rate from minimum value is plotted versus creep strain. The two solid lines represent the increase according to equation [13],

using values of 5 and 10 for n to represent the range observed for steels. The dotted curves represent actual values for an alloy steel and a carbon steel, computed from the White, Clark and Wilson (13) data. The upward inflection of creep rate is seen to progress at much higher rate than can be accounted for by reduction of area. Hence it becomes apparent that other factors must be sought to account for the acceleration of creep strain beyond the inflection for minimum creep rate. A time factor based upon intrinsic changes of quality

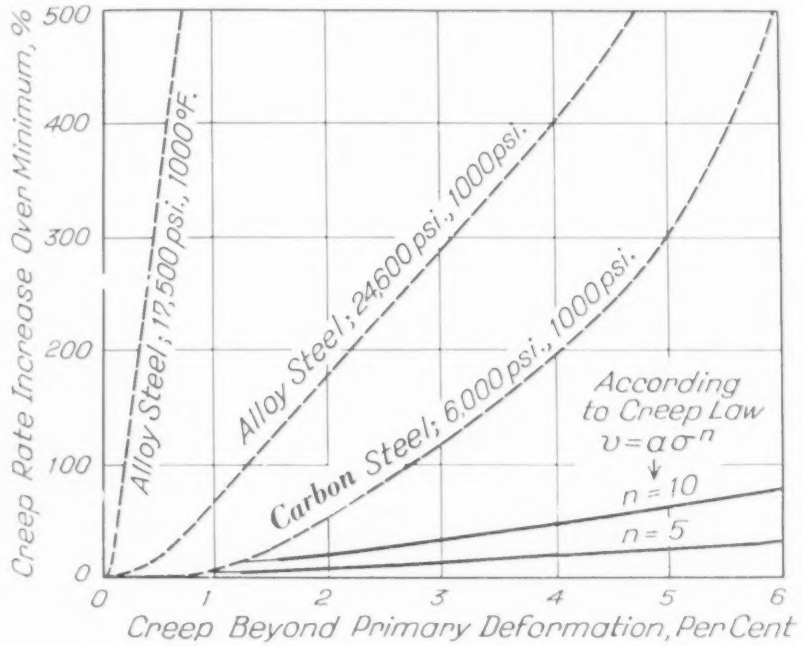

Fig. 3—Creep Rate Versus Creep Strain. Comparison of Actual Acceleration of Creep Rates with Acceleration Attributable to Reduction of Section.

which the material undergoes must be introduced in expressions for the creep strain-time relationship if extrapolations are to have any physical significance.

The evidence that creep deformation proceeds most readily along the crystal planes in best alignment with the planes of maximum shear, suggests the possibility of preferred orientation contributing to failure by creep. It is conceivable that several per cent of elongation under creep conditions is sufficient to rotate the crystal axes of a cubic metal to such a degree that each crystal presents potential slip lanes in better alignment with the planes of maximum shear than was the case originally. Such a realignment would indeed reduce

creep resistance and lend itself to rapidly accelerating strain subsequent to the minimum creep rate. Rotation of grains in creep has been experimentally established by Moore, Betty and Dollins (10). Microscopic examination was made of the surface of a bar of 2 per cent tin-lead alloy after stressing 1100 hours at 1000 pounds per square inch. No etching was necessary to bring out the grain structure

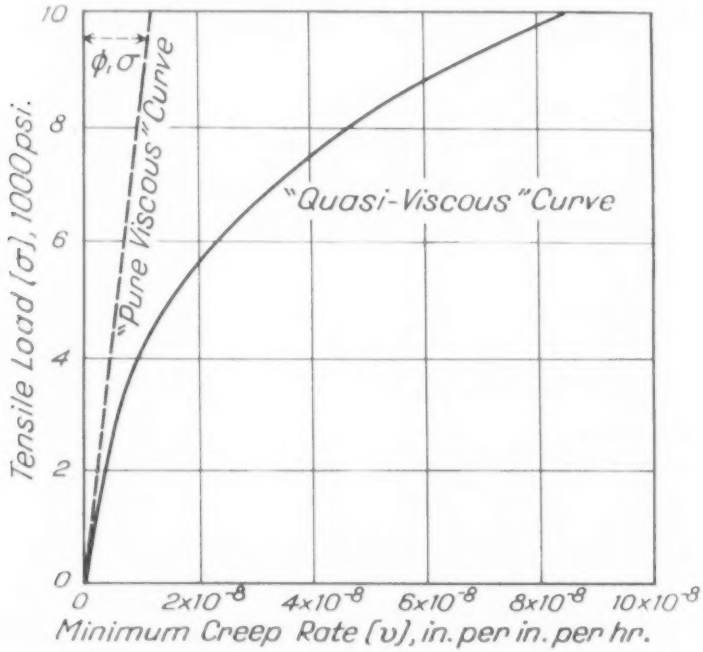

Fig. 4—Creep Rate Versus Stress According to the "Viscous" Interpretation.

of this alloy, because the grain boundaries were found to be clearly outlined by the shadows of raised edges. Microtome scratches left by the original sectioning of the surface were found to have become segmented, traversing each grain in a slightly altered direction. The ridges bounding each grain and the varied directions of the scratches, together, showed that rotation in space had occurred. Continuous recrystallization would seem necessary to produce rotation. Strain rates giving rise to strain-hardening at a rate too rapid to be offset by recrystallization would not be expected to permit clear cut rotation of grains to proceed. Short-time deformation is distinguished from creep deformation by just such a difference. Grain rotation should be expected to ultimately produce intergranular fracture, whereas under strain-hardening circumstances fracture should cause transverse cracks across the grains incurred by stress concen-

trations among crystallites. The manner of fracturing should therefore depend to a great extent upon the straining rate. Fractures produced in creep at a given temperature generally show much less reduction of area than in short-time tensile tests. Intergranular fracture frequently results in creep at temperatures giving transgranular fracture for short-time tensile testing. Recrystallization temperature and the "equi-cohesive" temperature would appear to be func-

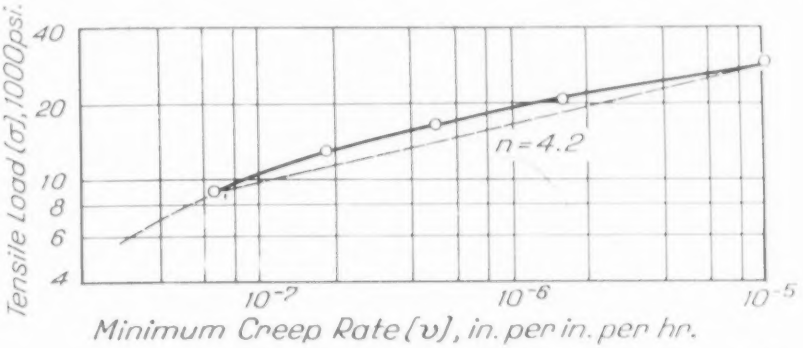

Fig. 5—Creep Rate Versus Stress. "Log-Log" Representation Used for Interpolation and Correction Purposes.

tions of the straining rate in accordance with the mechanism of deformation permitted by that straining rate.

Rate of Creep versus Stress

The dependence of stress in metals upon the rate of plastic straining has long been appreciated. This relationship was originally formulated by Ludwik (14) who observed that as the strain rate increased by geometric increments the dependent stress increased by arithmetic increments. This law can be written

$$\sigma - \sigma_o = \sigma_1 \log \frac{v}{v_1}, \qquad [14]$$

where $\sigma - \sigma_o$ is a range of stress attained under plastic deformation with σ_1 and v_1 constants. Creep testing further confirms the approximate validity of this law. In the creep test where the load is usually held constant, the creep rates resulting in a given material under different stresses are frequently correlated over a range extending from 10^{-8} inches per inch per hour to such rates as usually produced in a tensile testing machine. Such correlations are generally based upon minimum creep rates observed. This increasing of stress due to increase of strain rate is interpreted as another aspect of strain-hard-

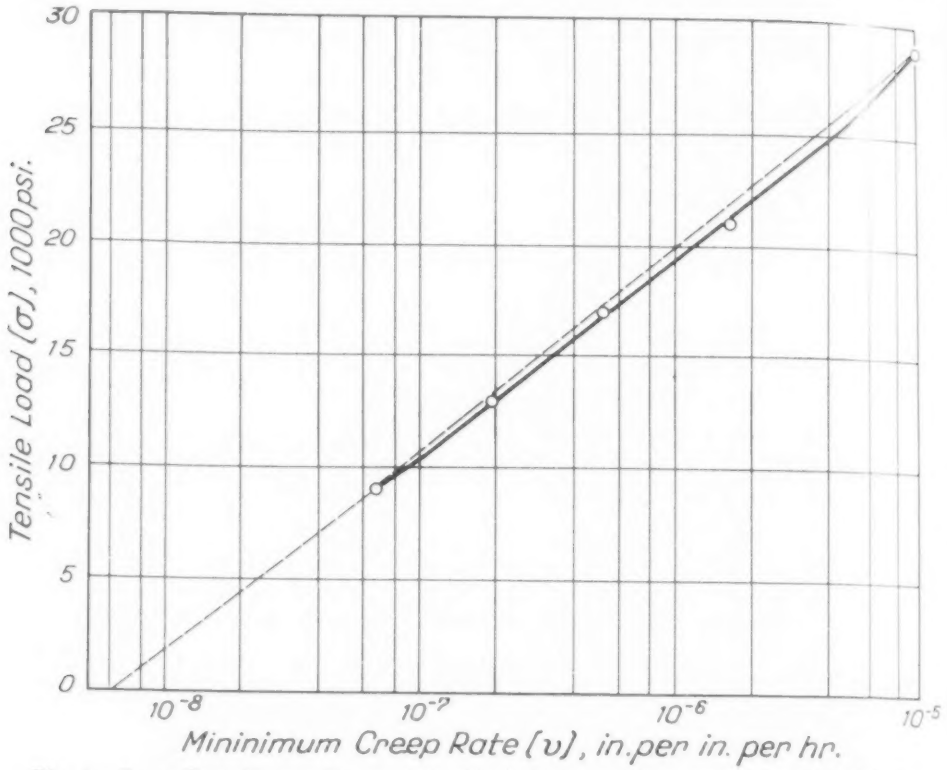

Fig. 6—Creep Rate Versus Stress—"Semi-log" Representation, Basis for Ludwik's Law.

ening, a different balance of hardening to annealing coming into equilibrium for each straining rate.

Ludwik's (14) relationship can be shown by semi-logarithmic plotting of data (Fig. 6), the stress rates being placed upon the log scale. A line drawn through data so plotted, when extrapolated toward lower rates of strain, intersects the zero stress ordinate at a finite strain value. This intercept apparently has no physical significance unless one assumes positive creep at zero stress. While such positive creep may exist in some materials at sufficiently high temperature, it seems more reasonable to believe there is a considerable departure from the Ludwik law for small stresses. Accurately measured creep values when plotted according to this law give slightly concave upward curve as illustrated by Fig. 6.

In order to rationalize the behavior of Ludwik's equations in the vicinity of very small strain rates, Soderberg (15) suggests a law satisfying the condition that $\sigma = 0$ for $v = 0$ expressed by the equation

$$\sigma = \sigma_1 \log \left(1 + \frac{v}{v_1} \right) . \qquad [15]$$

This creep law makes the tacit assumption that when v is very small the law becomes

$$\sigma = \frac{\sigma_1}{v_1} v = \phi_1 v. \qquad [16]$$

Viscous Creep

Nadai points out in discussion (15) that [19] is the expression for pure viscous flow. The quantity $\frac{\sigma_1}{v_1}$ has the dimensions of viscosity and is identical with ϕ. The practical importance of this interpretation is quite unique since it depends upon inferences which cannot as yet be satisfactorily explored experimentally. Hypothetical as it may be, in designing machines intended to operate for several decades without suffering appreciable permanent deformation, as for instance steam turbines, the role of viscous creep commands recognition. Soderberg (15) states that it is entirely possible that we shall have to be content with theoretical reasoning in tracing the creep relations for very low creep rates.

The creep rate-stress curve when plotted on straight co-ordinates (Fig. 4) is steeply inclined toward the stress ordinate at the lower stresses. Since considerable theoretical importance is attached to the shape of the initial portion of this curve by designing engineers concerned with minute creep strains, it is therefore essential that detailed consideration be given the relative merits of the extrapolation methods which can be used. Much minimum rate creep data has been presented using double logarithmic plotting. This scheme of representation shows a good straight line correlation in many instances. Such correlations imply power function of the stress, written

$$\frac{d\epsilon_p}{dt} = v = a\sigma^n. \qquad [17]$$

Extrapolation toward lower rates than measurable using a log-log plot is equivalent to assigning the initial stress a far smaller creep rate than would be the case were a semi-log plot similarly extrapolated.

Analyzing the most accurate available data for minimum creep

rates by the two methods of plotting it appears that the semi-log diagram correlates the data in better approximation of a straight line than does the log-log diagram which, as illustrated by Fig. 5, shows a considerable downward concavity. Minimum rates taken from published data on prolonged tests by Clark and Robinson (16) are given in Figs. 5 and 6 to illustrate the difference in character between the two types of diagrams.

There thus appears to be little justification for straight line extrapolation of log-log strain rate plots. A more reasonable method of extrapolation is one preserving the concavity of the determined portion of the curve. A better understanding of the implications of extrapolations on such logarithmic diagrams will be afforded through some discussion of the first derivatives of equations [15] and [17]. For [14],

$$\frac{dv}{d\sigma} = \frac{v}{\sigma_1} \qquad [18]$$

and for [15]

$$\frac{dv}{d\sigma} = na_1\sigma^{(n-1)} = a_2 v^{\left(\frac{n-1}{n}\right)}. \qquad [19]$$

The values of n determined for [17] or [19] from the slopes of log-log plots generally range from 5 to 10, depending on the temperature and material. Thus derivative [19] tells us that the acceleration of strain rate with respect to stress is proportional to the 0.8 to 0.9 power of the strain rate or just short of being linear. On the other hand derivative [18] states that the acceleration of strain increases in direct proportion to the strain rate and that as the rate of strain approaches zero, the strain rate and stress tend toward direct proportionality. At higher rates of strain equations [14], [15] and [16] can all be fitted to a range of data with good enough approximation for correcting experimental error (see Robinson 29), but it is quite obvious that straight line extrapolations of [14] and [17] on a basis of logarithmic plots has no physical significance for slow strain rates. Since the data show opposite directions of departure from laws [14] and [17] it is quite evident that the initial path of the true curve lies in a region bounded by the two curves according to [14] and [17]. Inspection of the power function derivative [18] reveals that any tangent to a curve on a log-log plot having unit slope, ($n = 1$), represents the equation of a straight line. Thus, if extrapolation of rate-stress data be made upon the log-log diagram pre-

serving the indicated curvature, a point is reached at which the tangent line determines a straight line representative of the initial stage of the rate-stress diagram. Although equation [15] gives a rational curve, a simpler function may be used, as will be shown, introducing the idea of viscous flow through more obvious and natural considerations.

Theoretical discussion of the creep rate-stress function is undoubtedly facilitated by introducing the concept of viscosity. Aside from the negligible effect of stress intensification resulting from constant density extension, all of the various factors which contribute their influence upon creep rate may be summed up in the instantaneous coefficient of viscosity, Φ. Thus the creep law is simply written as

$$\phi v = \sigma, \qquad [20]$$

Nadai (15) chooses to divide the rate-stress diagram (Fig. 4) into two ranges. In the initial range, creep is purely viscous following the law [16] and is completely described by the constant coefficient Φ, or the condition that

$$\frac{d\sigma}{dv} = \phi_1. \qquad [21]$$

The second range of creep distinguished as quasi viscous is characterized by a variable coefficient of viscosity which rises or falls with strain and time. Under a constant rate of straining, increase of unit stress corresponds to an increase of Φ. Such an increase may be regarded as an index or coefficient of the net strain-hardening. The change from the initial pure viscosity under this condition is written,

$$d\phi = \frac{1}{v_1} d\sigma. \qquad [22]$$

Similarly when unit stress is held constant, the decreasing creep rate of the primary stage signifies that the sum of strain hardening and time hardening is the dominating change and can be represented as by a rise in the viscosity coefficient. The creep law for constant unit stress is thus

$$v d\phi + \phi dv = 0. \qquad [23]$$

$$\frac{dv}{d\phi} = -\frac{v}{\phi}.$$

The condition for a secondary or minimum creep rate at constant stress in the creep-time curve is consequently satisfied when the rate of viscosity change due to the hardening influences is offset by an equivalent change due to the annealing effects. The final stage of creep is reached when the summation of annealing effects is sufficient to reduce ϕ.

To express the variable coefficient of *quasi* viscosity in terms of the various factors governing it would involve a function of staggering complexity. Such a function would be required at least to place the following factors in their correct interrelation.

(1)	Initial viscosity	(5)	Annealing
(2)	Work hardening	(6)	Orientation of grains
(3)	Age hardening	(7)	Strain energy
(4)	Work hardening relief	(8)	Temperature

The desirability of expressing $\frac{1}{\phi}$, the "flowability," directly in terms of the data suggests plotting its instantaneous values, $\frac{v}{\sigma}$, as a function of the stress σ. Such a plot is shown in Fig. 7 using the semi-logarithmic diagram to represent the same data given in Figs. 5 and 6. This type of plot gives a better straight line correlation of the minimum creep rate data than the Ludwik plot. The equation relating v and σ is thus

$$\log \phi_1 \frac{v}{\sigma} = k\sigma, \qquad [24]$$

or

$$\phi_1 v = \sigma e^{k\sigma}, \qquad [25]$$

where k is a constant. The behavior of this function when $\sigma = o$ implies pure viscosity, and that the viscosity decreases in geometrical proportion to the stress,

$$\phi = \phi_1 e^{-k\sigma}. \qquad [26]$$

The constant ϕ_1 has the dimensions of viscosity and is the pure viscosity coefficient. For steel at 857 degrees Fahr., extrapolating the data of Fig. 7 gives ϕ_1 the value of 8.3×10^{11}. The scatter of the points in Fig. 7 is such that those intermediate to the extremes lie on both sides of a line joining the extremes. In Figs. 5 and 6, the intermediate points all lie on the same side of the arcs joining the ex-

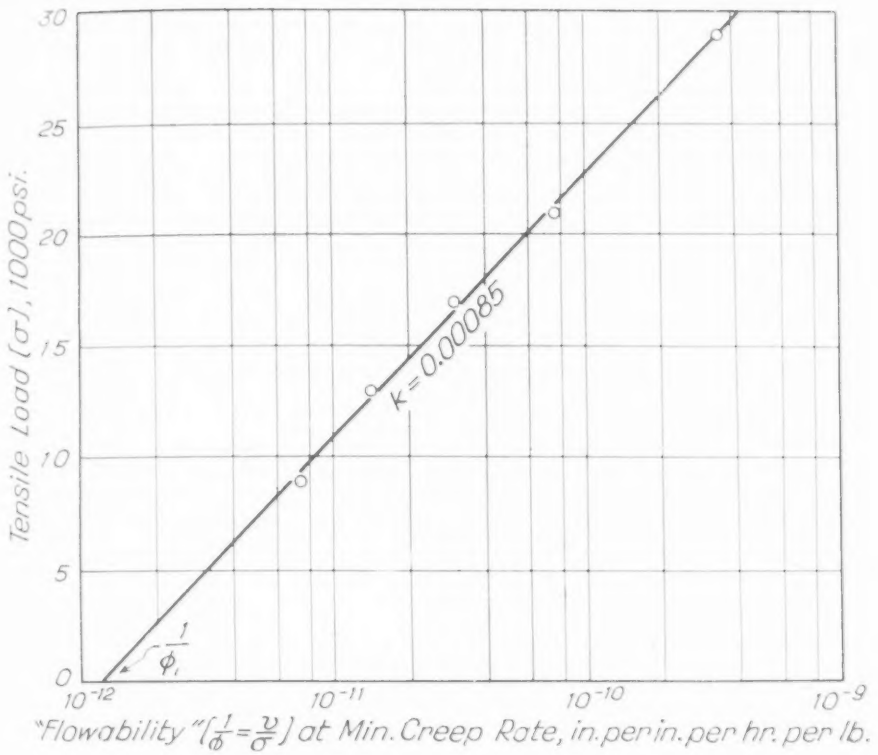

Fig. 7—Flowability (Reciprocal Viscosity) Versus Stress—"Semi-log" Representation, Basis for the Law of "Quasi-Viscous" Creep.

tremes, thus showing Fig. 7 to give a better straight line correlation than either Figs. 5 or 6. Treating the relationship between straining rate and stress as a viscosity function thus leads to a better correlation of creep rates than does Ludwik's law, and affords a more rational interpretation of creep at very small stresses.

The Theory of Creep by Shear

Combined Stress Experiments

The theory of creep deformation by shear has been developed largely through the work of Bailey (6), (12), (17). The nature of creep under systems of compound stress was deduced from a painstaking and ingenious series of studies upon the creep of thin walled tubes under various types of stress application, including

(1) Axial tension
(2) Torsion
(3) Combined torsion and axial tension
(4) Internal pressure
(5) Combined internal pressure and axial loading

Studies under conditions (1), (2) and (3) were accomplished using carbon steel tubes at temperatures of 900 degrees Fahr. (480 degrees Cent.) and 1020 degrees Fahr. (550 degrees Cent.). Due to the experimental difficulties in making high temperature creep tests on steel hydrostatically loaded, as encountered by White and Clark (18), Bailey confined his studies for conditions (4) and (5) to lead pipes at atmospheric temperature. All of Bailey's conclusions regarding combined stress relationships are based upon the assumption of uniform creep rate. The various tests made were permitted to continue until a uniform rate of creep strain could be estimated. Elaborate precautions were taken in making comparisons to rule out effects due to work hardening. In the series of tests representing combined torsion and axial tension, steel tubes were machined from bar stock to obtain a better approach to isotropic specimens.

Tubes under torsion, which is the equivalent of pure shear, attained a uniform rate of torque, but showed no measurable axial deformation. This fact is of prime importance since it permits the deduction to be made that creep by shear is practically uninfluenced by the stress acting at right angles to the plane of shear. This is made clear if the shear stress is resolved into its equivalent principal tensile and compressive stresses acting upon planes at 45 degrees to the axis of torque. Since the creep arising from shear on these planes should be uniform, no creep in the direction of the axis is to be expected. Therefore, as a working hypothesis, Bailey treats the influence of stress acting at right angles to a plane as being immaterial, whether it be tensile or compressive. As a corollary of this hypothesis it follows that creep is dependent only upon shear stress. Moreover, creep under simple tension and simple compression should be of equivalent magnitude.

A thin walled tube under internal pressure is influenced by a three dimensional system of stress. The circumferential or "hoop stress," σ of such a system is twice the axial stress. The radial stress, which is of negligible magnitude, varies from the unit pressure on the inside of the tube to zero on the outside. For practical purposes the circumferential and axial principal stress only need be considered. Bailey (12) finds that thin tubes under normal hydrostatic internal pressure creep circumferentially with no measurable axial deformation. If, however, the normal two to one ratio of "hoop" to axial stress is disturbed, though superimposing an axial stress, the circumferential strain is found to increase. For either increase or de-

crease of the normal ratio between the two principal stresses, axial creep in the direction of the superimposed stress takes place. The lead pipes studied crept five times as rapidly, circumferentially, under simple "hoop" stress, as when under normal internal pressure conditions. Combining "hoop" stress, σ with axial stress twice as great, 2σ, produces rapid axial elongation, but no circumferential creep, although failure occurs by circumferential fracture. Tests thus indicate that a thin wall tube under internal pressure without any superimposed axial loading is under the most favorable conditions of two dimensional stressing for creep resistance.

Bailey's Interpretation of Experiments

The modes of deformation for thin tubes under torsional creep and internal pressure have been established as similar in nature. In fact Bailey suggests that a torsional creep test provides a very ready means of deducing the creep behavior of a tube under internal fluid pressure. This idea has been justified by Robert's (17) demonstration that like resolved shearing stresses, whether produced in torsion or by internal pressure, give rise to similar creep strain rates. The identity of the two systems of stresses with regard to shear resolution may be shown by the artifice of superimposing a hydrostatic tensile stress upon the torsional stress system. Such a three-dimensional stress, in which the principal stresses are all equal and positive, could not give rise to creep by shear since deformation could take place only by increase of volume. Thus addition or subtraction of such a hydrostatic tensile stress may be used to transform one two-dimensional system into another which is its equivalent so far as deformation by shear is concerned. The transformation applying to the shear stress conversion of cylindrical torsion to cylindrical internal pressure is illustrated in Fig. 8 where the two systems can be seen to differ by a hydrostatic tensile stress of $\frac{\sigma}{2}$.

When the creep rates for a given tube acting in the direction of major principal stresses are compared for the above systems it is found that the pure tensile creep rate for a given value of maximum shear is greater than the torsional creep rate in the direction of principal stresses for the same value of maximum shear. In the case of steel tubes at 900 degrees Fahr. (480 degrees Cent.), Bailey finds

the ratio ranging from 1.9 to 2.3, while at 1020 degrees Fahr. (550 degrees Cent.) the ratio range is 2.7 to 3.2. In the case of internal pressure, a ratio of similar significance is observed as the diametral creep rate increases upon neutralization of axial stress. Bailey's lead pipe experiments fixed the value of this ratio for lead at about 5. Bailey (6) argues that this effect is partially accounted for by considering the three stress systems depicted in Figs. 9a and 9b. The

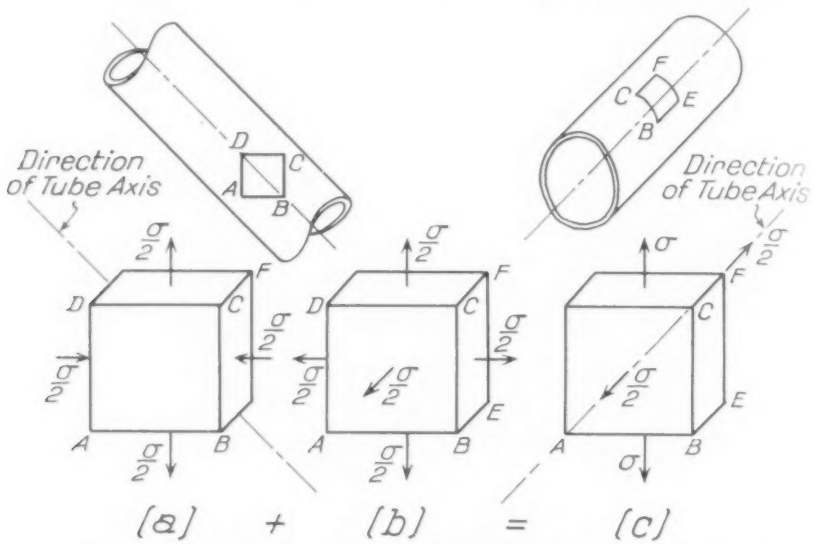

Fig. 8—Similarity Between the Maximum Shear Stress Systems Influencing the Creep of Thin-Walled Tubes in Torsion (a) and Under Internal Pressure (c). Hydrostatic Tensile System (b), Having no Creep. When Superimposed Upon (a) Converts (a) to (c). Bailey (12).

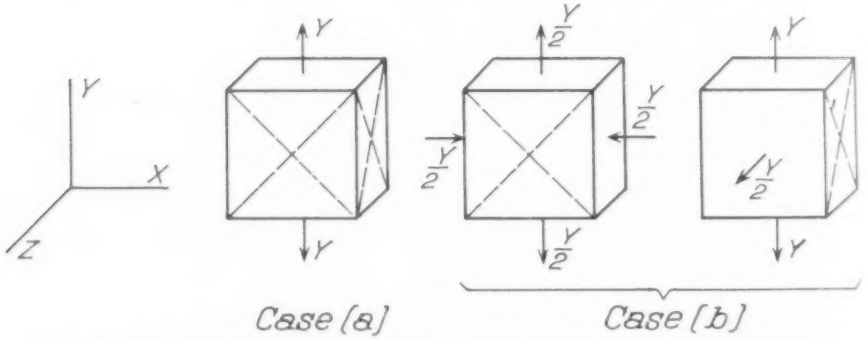

Fig. 9—Directions of Maximum Shear Under Simple and Combined Stressing. Bailey (6).

shear stress on all planes parallel to the Z-axis is the same in both cases. In the case (a) the shear stress on planes parallel to the X-axis is similar to that on corresponding planes parallel to the Z-axis, which with isotropic material would be expected to contribute equally to creep in the direction of the Y-axis. The stress systems (b)

which have already been shown equivalent in creep, having shear stresses on the planes parallel to X-axis equal to half of that on the corresponding planes of case (a), should creep less in the direction of the stress. The view that creep depends only upon shear stresses acting respectively on each of these sets of planes requires that creep in the direction of Y, case (a) shall not exceed twice that in case (b). Since this ratio is actually exceeded, as found by experiment, it seems that the shear stresses on the two sets of planes considered are not independent in their promotion of creep. It would appear, therefore, that a general system of shear stresses exists, acting on planes parallel to all three axes and capable of influencing the creep arising from shear upon planes parallel to one or other of the two axes normal to the direction of creep.

A generalized expression of creep by shear at constant temperature on these principles has been attempted by Bailey (6), stated in terms of the three principal stresses. Creep in the direction of a principal stress he represents by one factor depending upon the planes where shear contributes to creep in the direction considered, and a second depending on the stress system as a whole, writing

$$v_x = f(X - Y) + f(X - Z). \qquad [27]$$

where X, Y and Z are the principal stresses and v_x, v_y, and v_z the respective creep rates in the directions of these stresses. This he states follows if creep is equally dependent upon the shear stresses in the two sets of planes parallel to the two directions normal to which there is no creep. Similarly the creep rate when applying to the other two principal directions would be respectively

$$v_y = f(Y - Z) + f(Y - X). \qquad [28]$$

$$v_z = f(Z - X) + f(Z - Y). \qquad [29]$$

Since constant density persists, the function must satisfy the condition that

$$v_x + v_y + v_z = 0. \qquad [30]$$

It is necessary therefore that the shear stress functions be such

$$v_x = f(X - Y) - f(Z - X). \qquad [27a]$$

$$v_y = f(Y - Z) - f(X - Y). \qquad [28a]$$

$$v_z = f(Z - X) - f(Y - Z). \qquad [29a]$$

This relation is furthermore required since it is necessary for the experimental fact, that in a system consisting of the two principal stresses $X = \frac{1}{2}Y$, that $v_x = o$ be obeyed. Bailey uses the power function [12] with the reservation that the magnitude of n does influence the signs of the stress, obtaining

$$v_y = a[(Y - Z)^n + (Y - X)^n]. \qquad [31]$$
$$= a[(Y - Z)^n - (X - Y)^n].$$

Bailey shows the ratio between the creep rates in the cases (a) and (b) Fig. 9 as becoming

$$\frac{v_{x=o}}{v_{x=\frac{1}{2}y}} = \frac{2Y^n}{\left(\frac{Y}{2}\right)^n + Y^n} = \frac{2}{1 + (\frac{1}{2})^n}. \qquad [32]$$

With experimental values of n varying from 5 to 10, this ratio approaches 2 very closely but cannot exceed this value and becomes 1½ at the other extreme where for viscous creep n = 1.

The inadequacy of Bailey's general expression, so interpreted, in explaining the experimental facts is at once apparent if we consider the experiment with lead pipe under internal pressure where a minimum creep was found at $X = \frac{1}{2}Y$ in 1:5 ratio to the creep at $X = o$. When X was made equal to Y this ratio became about 1:2½, while at $Y = \frac{1}{2}X$ it is diminished to 1:½. These ratios according to expression [31] may be summarized as follows

	Ratio	Experimental	Bailey's Expression
(1)	$\frac{v_{x=o}}{v_{x=\frac{1}{2}y}}$	5, 3, 2	$\frac{2}{1 + (\frac{1}{2})^n} = 1.97$, approx.
(2)	$\frac{v_{x=o}}{v_{x=y}}$	2½	2
(3)	$\frac{v_{x=o}}{v_{x=2y}}$	½	∞

If in case (3) signs are given the same interpretation as Bailey places upon them in case (1), the ratio becomes infinite.

In an attempt to bolster up the quantitative agreement of the ratio in case (1) with experiment, Bailey introduces specific shear strain energy. Under creep conditions he points out an appreciable time interval must elapse between the occurrence of individual shear

slips. It appears reasonable therefore to consider that the work done during each increment of deformation is obtained at the expense of the available potential energy. That the strain energy in the case of simple tension or compression is proportional to the square of the stress follows from Hooke's law, $\sigma = \epsilon E$, since the work of elastic deformation is

$$W = \epsilon\sigma = \frac{\sigma^2}{E}, \qquad [33]$$

where E is Young's modulus. The specific elastic shear-strain energy for a general stress system has been shown by Hencky to be proportional to

$$\tfrac{1}{2}(X - Y)^2 + \tfrac{1}{2}(Z - X)^2 + \tfrac{1}{2}(X - Z)^2. \qquad [34]$$

Shear strain energy is applied to the general creep expression as a factor involved in the coefficient a upon the argument that resistance to slip is lowered through local heating caused by the release of shear strain energy and that this lowering might be considered proportional to a power m of the shear strain energy. Where $n = 2m + q$ the expression for simple tension then becomes

$$v = A\,\sigma^{2m+q}. \qquad [35]$$

Modifying the general expression accordingly and applying it to the ratio of cases (a) and (b) Fig. 9 gives

$$\frac{v_{x=0}}{v_{x=\frac{1}{2}y}} = \frac{2}{(\tfrac{3}{4})^m\,[1 + \tfrac{1}{2}^{(m-2m)}]}. \qquad [36]$$

Where $m = 1$, as would be the case, if the decrease in slip resistance were considered directly proportional to the strain energy, the ratio assumes the value 2.67, provided $n = 6$. In order to have a ratio of 5, as found for lead, m must be assigned a value of 3 or more.

Relationship Between Elasticity and Creep

Despite corrections for shear strain energy, the general expression applies only to case (1), while for other cases refinement of the hypothesis regarding effects of shear stresses upon the system as a whole is necessary if a rational interpretation is to be attained.

The idealized conception of creep by shear tacitly assumes, as a consequence of the constant density relationship, that when stress is applied in one principal direction the stresses acting in the other two principal directions remain zero. While such would be true of a perfectly isotropic and plastic substance, metals under creep do not fulfill this condition. At creep strain rates of small magnitude slip on a plane is occurring at discrete intervals of atomic pitch corresponding roughly to once an hour for rate of 10^{-5}, once in 4 seconds for 10^{-8} (10 per cent a year). The instantaneous condition of the metal is therefore one of essentially elastic strain, with occasional disturbances due to slip. Experimentally, the elastic contraction after creep is equivalent to that prior to creep. The actual frequency of the disturbances upon any given plane is probably far less than the above estimate, for when slip does occur, it is attended by strain hardening and the formation of crystallites, indicating block movements of many atomic pitches. A limited region of the material, it would seem, is in a state of elastic strain practically all of the time, subject to continual elastic tremors of varying intensities and periodicities, initiated by the slip in neighboring regions. We might conceive of a piece of creeping metal as an earthquake region. The event of slip is a local cataclysm, disturbing neighboring regions in a degree proportional to their proximity. Each cataclysmic change causes a slight redistribution of stress, of which the accumulations serve to bring about further slip cataclysms. The creep recovery which occurs upon release of stress might be attributed in part to continuation of this kinematical activity through the dissipation of unequally distributed local stresses, resulting in a reversal of block slip.

Although creep by shearing stress has a kinematic aspect, a picture of the static state of elastic stress in a metal is helpful in interpreting creep results. When a unidirectional stress is applied to a metal, a three-dimensional system of internal stress is set up. The creep rate which we measure results from the shear stress in such a system. Poisson's ratio, λ, lateral unit contraction divided by unit elastic elongation, furnishes something of an index of the intensity of this field of internal strain. If elastic elongation of a material were accompanied by no change of volume, the ratio λ would be 0.5 and we could conclude that all of the strain energy and force were applying in the direction of elongation. But metals all have λ values less than 0.5, ranging from 0.17 for hard material like tungsten to 0.43

for soft material like lead. Steel when cold has a value of 0.3 which probably increases to about 0.4 at temperatures where creep is studied. Elastic extension is accompanied by an increase of volume which we may consider gives rise to equivalent principal force components in a polycrystalline metal in the plane normal to the applied force. The magnitude of each of these components is roughly, $(\frac{1}{2}-\lambda)\sigma$. That is to say if a material has a Poisson's ratio of 0.40, the lateral stress components will be of the order of 10 per cent of the applied tensile stress. The force acting upon the planes of maximum shear in tension will be in this case about 45 per cent of σ instead of 50 per cent as conventionally computed. Now a 10 per cent increment of stress decrease is enough to effect about 45 per cent decrease of creep rate.

If such a considerable effect upon creep rate can be attributed to the Poisson ratio for tension, the question arises whether one is justified in assuming the compressional Poisson's ratio equivalent to the tensional. Even a slight difference between the two values would account for an appreciable difference between tensional and compression creep rates under the same applied stress. This consideration may afford an explanation for the creep ratio between torsional stressing and internal pressure stressing as observed by Bailey. These experiments tended to show creep under torsional shear proceeding 2 to 3 times faster than for internal pressure producing an equivalent shear stress. When it is considered that such a ratio in rates could be effected by about a 10 per cent change in stress, one is led to inquire to what extent discrepancies in resolved shear stresses attributable to Poisson's ratio and testing technique may have influenced the creep rate ratios as determined by Bailey. Torsional stressing involves a compressive component and therefore is relatively more compressive to a material than internal pressure on thin wall tubing wherein the principal stress components are tensional. If a proper correction were made for this effect, the results from the torsion tests might be brought into better accord with the internal pressure results.

Summing up the various considerations given the mechanism of creep, there seems to be justification for setting up the general creep rate expression as a function of the average maximum shear stresses. Doing so amounts to recognition of the 5 to 1 creep rate ratio for the internal pressure condition in the lead experiments as meaning that the effective maximum shear stress is about 25 per cent less

under the combined stressing than for simple tension. The usable creep stress for materials in a closed end tubular pressure vessel of lead is thus 33 per cent greater than the creep stress derived for simple tension. The general creep rate expression according to this interpretation should be written,

$$v_x = f(2X - Y - Z). \qquad [37]$$

Whereas the additive scheme of expression, equation [27], assumes a discrete identity of the various maximum shear stress components, equation [37] gives more weight to the idea that the creep under any combination of stress is the resultant of slip by shear in all of the planes which intersect the creep axis. Moreover, the expression in this form needs no further assumption for evaluating internal pressure creep than the hypothesis that the exponential relationship between creep rate and stress is approximately the same for combined and simple stresses.

The conception of viscous creep at small stresses suggests a somewhat different choice of function for the general creep expression than used by Bailey. These equations assume a much simpler and rational form for all values of stress when expressed in terms of a viscosity function, becoming

$$\phi_{v_x} = (2X - Y - Z). \qquad [38]$$

This form of the general expression has the immediate advantage that the constant density relationship is implied without a special rule concerning the signs of principal stress components as necessary when the power function is used.

Introducing the stress function of viscosity, the general creep expression in terms of principal stresses becomes

$$\log (\phi_1 v_x) - \log \tfrac{1}{2}(2X - Y - Z) = \frac{\kappa}{2}(2X - Y - Z). \qquad [39]$$

$$\log (\phi_1 v_y) - \log \tfrac{1}{2}(2Y - Z - X) = \frac{\kappa}{2}(2Y - Z - X). \qquad [40]$$

$$\log (\phi_1 v_z) - \log \tfrac{1}{2}(2Z - X - Y) = \frac{\kappa}{2}(2Z - X - Y). \qquad [41]$$

Considering again the cases (a) and (b) Fig. 9 in the light of the viscosity relationship, the ratio is

$$\frac{V_{x=0}}{V_{x=\frac{1}{2}y}} = 4/3\, e^{\frac{\kappa y}{4}}. \qquad [42]$$

The ratio is thus linked with stress intensity in a simple manner. The pure viscous creep at small stresses is associated with a ratio of 1.33. This implies that as internal pressure rises, however, the 33 per cent strengthening is more effective in reducing the diametral creep. The coefficient κ for carbon steel at a temperature of 850 degrees Fahr. (455 degrees Cent.) has a value of about 4×10^{-4} reciprocal pounds. Using this value with $\phi_1 = 18.3 \times 10^{10}$, the following table is computed.

Simple Tension Creep Stress*	Creep Rate in Tension	$\frac{V_{x=0}}{V_{x=\frac{1}{2}y}}$	"Hoop" Stress for Equivalent Creep
0	0	1.33	0
1000	0.8×10^{-8}	1.47	1333
3000	6.0×10^{-8}	1.80	4000
10000	310.0×10^{-8}	3.62	13333

*Lbs. per sq. inch.

As temperature increases larger values of κ are found, carbon steel at 1000 degrees Fahr. (540 degrees Cent.) having a value of about 10^{-3} reciprocal pounds, with $\phi_1 = 14 \times 10^{10}$ approximately. The internal pressure calculations for this temperature are as follows.

Simple Tension Creep Stress*	Creep Rate in Tension	$\frac{V_{x=0}}{V_{x=\frac{1}{2}y}}$	"Hoop" Stress for Equivalent Creep
0	0	1.333	0
100	0.08×10^{-8}	1.475	133
500	0.59×10^{-8}	2.19	667
1000	1.93×10^{-8}	3.60	1333
2000	10.35×10^{-8}	9.65	2667
4000	150.8×10^{-8}	70.4	5333

*Lbs. per sq. inch.

This somewhat astonishing calculation, showing the relative reduction in creep rate due to pressure strengthening to be considerably more at 1000 degrees Fahr. (540 degrees Cent.) than at 850 degrees Fahr. (455 degrees Cent.) is substantiated by actual field experience. Measurements of creep rate made upon tubes in actual service show in general that the rates are less than would be anticipated from laboratory tensile creep tests. Super heater tubes and oil still tubes are used under conditions where rates of the magnitude 1 per cent a year are tolerable. Failure in such services more often occurs through corrosion and scaling than creep deformation.

Consequently, it is desirable to stress the tubing for all the deformation it will stand and some engineers have not hesitated to disregard laboratory creep results as too conservative. The laboratory creep results apparently are not too conservative for computing tube stresses if a proper interpretation of their application is made.

Creep Under Transverse Loading

Creep in bending or transverse deflection is usually treated upon the hypothesis that for any cross-sectional form the initial moment plane rotates about the neutral axis without developing curvature. This follows as a necessary condition for constant density. The rate of creep at any point in the cross section is consequently proportional to the distance from the neutral axis. Bending tests by McCullough (19) and by Tapsell and Johnson (20) substantiate this conclusion. The fibre stress, therefore, does not vary linearly from the neutral axis to the surface as it would under elastic conditions, but in accordance with the creep law. The more highly viscous creep in the vicinity of the neutral axis offers comparatively more resistance to deformation than the less viscous creep near the outer fibre. This results in a more nearly uniform distribution of stress than for elastic deformation, reducing the maximum fibre stress. The moment of resistance is generally taken to be about 0.9 of that for uniform stress.

The Relaxation Problem

The relaxation of stress in metal under a fixed or limited strain is an important case of creep encountered in practice. Bolts, studs, flanges and springs of any kind in high temperature applications are subject to this kind of creep. The loading of such member depends upon the maintenance of an elastic strain; consequently as creep occurs the load diminishes. Usually two or more parts are coupled in series so that the decrease of stress for a given amount of creep is governed by the rigidity of the system as a whole. The problem involved is to relate residual stress and time. A machine for studying relaxation of stress by creep was devised by Barr and Bardgett (21). Their device permits the measurement of residual stress during a test in terms of the elastic deformation in portion of the system not subject to creep. This is accomplished by means of a cold "weigh bar" or extensometer in series with the heated creep test bar. A typ-

ical record of stress versus time in a test by this scheme is shown in Fig. 10.

It is obvious that the residual stress in the system will be proportional to the difference between the initial elastic deformation and the accumulated creep. If a perfectly rigid machine could be built for testing a bar, the expression for residual unit stress σ_r in terms of Young's modulus would be

$$\sigma_r = E(\epsilon_e - \epsilon_p). \qquad [43]$$

Practical cases, however, always involve a total elastic displacement greater than that of the creeping portion of the system, as though a

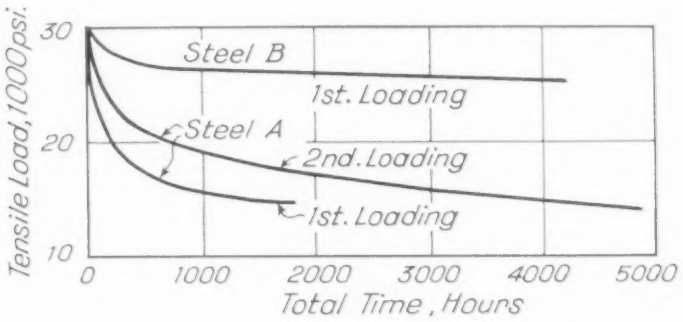

Fig. 10—Typical Relaxation Test Curves—Steel A Relaxes Less Upon Second Loading Due to Work-Hardening. Steel B Has Less Primary Creep than Steel A, Accounting for its Smaller Initial Relaxation. Secondary Rates of Relaxations Differ Showing Steel B to Have Greater Creep Resistance to Steel A.

spring were interposed in the perfectly rigid system. In the case where the creeping member has a uniform section of length, l, and the total elastic displacement of the system is h times ϵ_e, the expression becomes

$$\sigma_r = E\left(\epsilon_e - \frac{1}{h}\epsilon_p\right). \qquad [44]$$

Certain difficulties arise when one attempts to solve this equation for time. Total creep, as has been discussed at length, is affected by strain hardening. The relaxation creep is not only affected by strain-hardening but is also influenced by diminishing stress. Baumann (22) in his treatment of the problem neglects the effect of strain-hardening and introduces an expression for constant creep rate as a function of stress. If the power function [12] be used the following expression is obtained

$$\frac{d\epsilon_p}{dt} = aE^n \left(\epsilon_e - \frac{1}{h}\epsilon_p \right)^n, \qquad [45]$$

which upon integration becomes

$$t = \frac{h\left[\left(\epsilon_e - \frac{1}{h}\epsilon_p\right)^{1-n} - \epsilon_e{}^{1-n}\right]}{al(n-1)E^n}. \qquad [46]$$

This equation can be used with a fair degree of approximation for computing relaxation curves provided the constants a and n are chosen to represent the phase of creep influencing the relaxing member. At the first stressing, a bolt for instance relaxes largely due to primary stage creep. The constants a and n therefore when applying to a computation of relaxation at first stressing, should be determined by the average primary creep rate. Successive stressing is known to lower the relaxation rate as a consequence of strain-hardening. Secondary creep rates are to be used in such computations. The necessary data for a and n can be conveniently evaluated from creep rates determined in a "stepdown" loading test in which the total creep at any stage is adjusted to simulate that incurred by the relaxing member.

Soderberg (15) developed a method of relaxation analysis in which strain-hardening is taken into account. This is accomplished by adding a strain-hardening term to the equation relating stress and rate of creep,

$$\sigma = f_1(v) + f_2(\epsilon_p). \qquad [47]$$

This correction term is introduced through a coefficient of work hardening, $E_p = \frac{d\sigma}{d\epsilon_p}$. Although experiment does not indicate an actually constant E_p, the error introduced by the simplification is not important in the treatment of relaxation. The equation used by Soderberg is as follows

$$\sigma = \sigma_1 \log \left(1 + \frac{1}{v_1}\frac{d\epsilon_p}{dt} \right) + E_p\epsilon_p. \qquad [48]$$

This equation is obviously valid only for deriving the constants pertaining to relaxation under primary creep at a decreasing rate, but, is useful in the detailed analysis of relaxation where decreasing pri-

mary creep is the principal consideration. The application of this function in equation [44] cannot be made, however, by straightforward procedure, the introduction of stress and time parameters being necessary.

Relaxation of Bolts

The amount of relaxation for a given total creep is inversely proportional to h. Hence any artifice used to increase h will effectively reduce the rate of relaxation. This principle has been used by McCutchan (25) to reduce the relaxation of studs in pipe flanges, at temperatures where difficulties in keeping tight joints due to creep are encountered. By interposing U-springs between the nuts and the flange, McCutchan increased the life of joints subjected to 1000-1100 degrees Fahr. (540-595 degrees Cent.) steam to sixteen times normal. Various other schemes incorporating this same principle suggest themselves as means of "side tracking" the relaxation problem, enabling the successful application of high elastic limit steels of low creep resistance.

Where h is limited to a low value and cannot conveniently be increased, materials with small initial creep are desirable. Various schemes have been suggested for "seasoning" bolting material to attain this end. Roberts' discussion (22) suggests giving steel bolting material a preliminary stretch under the estimated working load at a temperature somewhat in excess of its operating temperature and cooling to atmospheric temperature with the *load still on*. This treatment should put the material in condition to creep at its minimum rate provided the bolt is loaded or pulled before reheating. Such creep seasoning would obviously be involved, expensive, and perhaps impractical. Much effort has therefore been expended in finding steel that will respond to heat treatments that will give adequate elasticity and yet render them in good condition to resist relaxation. There are certain limitations on what can be accomplished in this regard because the treatments which afford the required elasticity seem to promote initial creep. If a tempered steel with a highly dispersed carbide structure is employed, the agglomeration and spheroidization ensuing at operating temperatures promotes primary creep. High elasticity attained through cold working is still more undesirable at creep temperature, since it is disposed to show even higher primary creep than for tempered steel. The choice of alloy steels which will retain adequate strength and elasticity when a properly agglomerated

carbide structure is produced is rather limited. Since the metallurgical aspects of this phase of the creep problem are beyond the scope of this paper, a detailed discussion of the various types of relaxation resistant steels, with their treatments, and resulting properties and structures, is not given.

The relaxation requirement of steels and their comparative per-

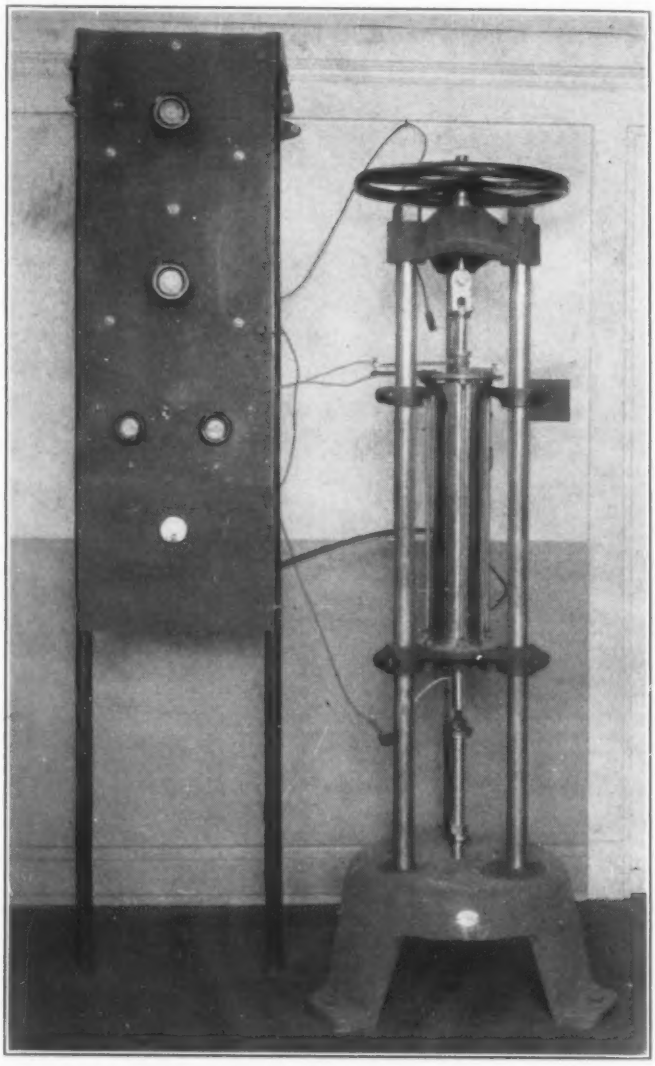

Fig. 11—Relaxation Testing Unit.

formances have not only been studied through mathematical analysis of constant load data, but through tests using the Barr and Bardgett scheme. A machine being used at Crane Co.'s Metallurgical Labora-

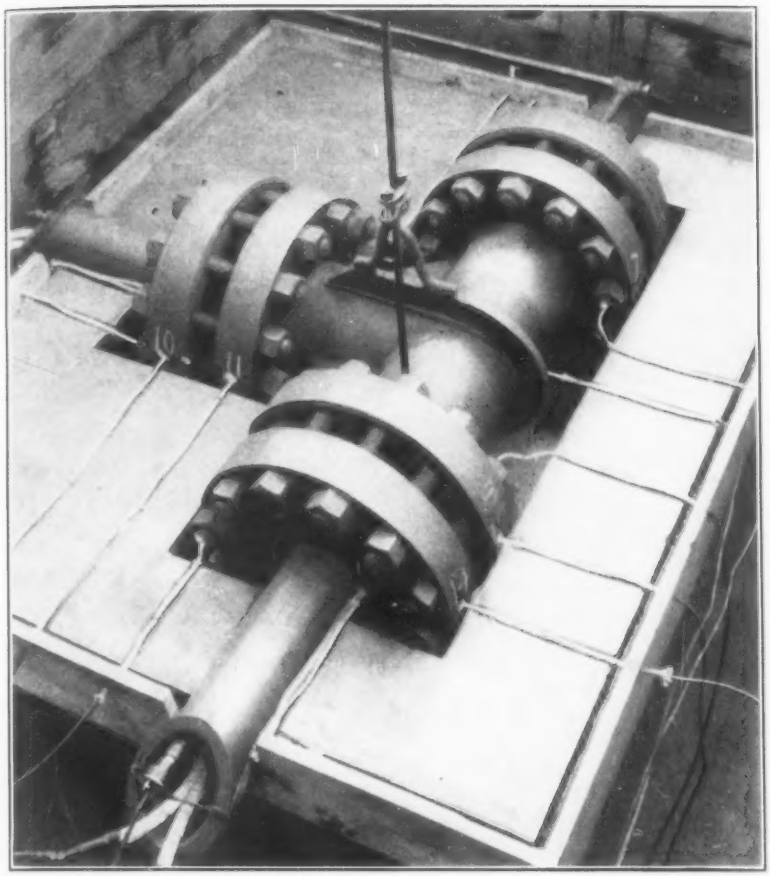

Fig. 12—Steel Flanged Tee Fitted for Studying Relaxation of Bolts Under High Pressure and Temperature Working Conditions.

tory for this purpose is shown in Fig. 11. Elastic deformations of the "weigh bar" in this machine are measured by means of a micrometer screw arrangement and temperature is held constant by maintaining the thermal expansion constant in an 18-8 steel sleeve liner in the tubular furnace. When using a 0.505-inch diameter test bar with a 6-inch long relaxation section, this machine has an h factor of 30. By using different length relaxation sections and various diameter test bars, this factor can be adjusted to simulate any specific service application and thereby produce a relaxation curve comparable to those expected in service.

The combined relaxation of a bolt and pipe flange system under internal pressure has been studied by Gough (23). A test model using internal unit heaters and means of controlling the steam pressures and temperatures was devised. Sensitive optical means of

measuring flange and bolt deformation were employed in recording the relaxation taking place in tests at steam temperatures of 950 degrees Fahr. (510 degrees Cent.) with 1400 pounds per square inch in pressure. Data was thus obtained for the creep of the flanges, gaskets, and nuts, as well as that in the bolt.

Creep measurements were made upon a 10-inch welded steel flanged tee under test conditions by Crane Co.'s Research Testing Laboratories. Internal pressure and temperatures were attained us-

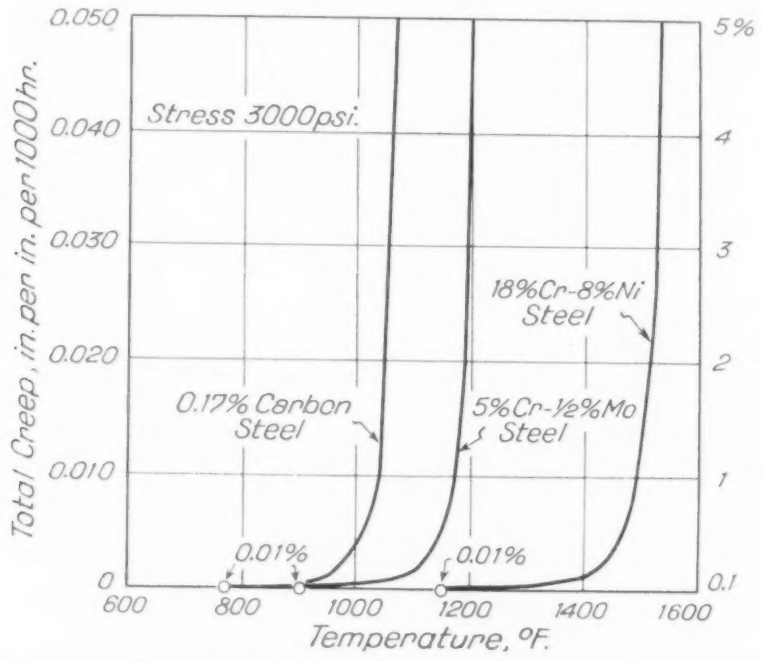

Fig. 13—Dependence of Total Creep Upon Temperature. Comparison of Steels Under 3000 Pounds Per Square Inch Stress Upon a 10,000-Hour Basis.

ing unit heaters and a pressure control system was provided. As a safeguard against unexpected bursting, the test fitting was situated in a pit and covered. The photograph (Fig. 12) shows the fitting, *in situ* with the upper covering removed. Temperatures were recorded at a number of points both internally and externally as shown by the thermocouple wiring. Increase in the diameter of the welded structure due to creep was determined by dial indicator measurements. Tests were run until the relaxation of the bolts and flanges caused leakage at which point the residual stresses in the bolting were determined. Direct comparison on the relative merits of various bolting materials was made by using a set of different type in each of the three flanges. Tests were made up to 1200 degrees Fahr. (650

degrees Cent.) at 1500 pounds per square inch internal pressure. Initial stud stresses of 30,000 pounds per square inch were used. It was established in these tests that leakage occurred in most instances when the bolt stress had relaxed to 15,000 pounds per square inch. At this stress the total bolt tension very nearly balances the hydraulic load upon the gasket area. These tests were discontinued when failure of the carbon steel welded tee by creep ultimately occurred at 1200 degrees Fahr. (650 degrees Cent.). The stresses in this fitting were estimated as 10,000 pounds per square inch—far in excess of permissible A.S.M.E Boiler Code stress which limits this grade of material to 4000 pounds per square inch at 950 degrees Fahr. (510 degrees Cent.).

The Use of Creep Tests for Determining the Relative Limiting Service Temperatures for Metals

The metal stress to be used in designing a structure for high temperature is frequently fixed by the necessity for metal thickness, shape or dimensions having to conform to established practices. Such being the situation, little latitude is afforded for adapting the design structure to the creep properties of a specific material. If consideration is given to creep the question hence arises—what will be the creep behavior of the available materials at the particular stress concentration in question? How will these materials behave at such stress as the temperature varies, and what are the maximum temperatures they will withstand without creep becoming objectionable?

The temperature maximum at which the required stress is tolerable is important, for if this temperature be exceeded by no more than 50 degrees Fahr., the rate of creep may increase several fold. To safeguard against the possibility of so seriously exceeding working limits, it is essential to evaluate the temperature-creep relations for the stresses in question.

Most of the creep investigation and studies which have been made focus attention on the stress-creep relationship at one or more picked temperatures. In such investigations the aim is to establish the precise stress yielding not more than the limiting creep rate. It is often difficult to estimate accurately from this data, accumulated from the stress-creep standpoint, the extent to which the creep rate changes with temperature and moreover to pick out the range of

temperature in which the metal is transformed to a weak, viscous character.

It is convenient and expedient in gaining creep information, as to the relative temperature limitations of materials in a given application, to conduct tests with temperature as the independent variable. In this manner the suitability of a material at a usable stress can be determined directly. Experimentally the procedure resolves itself into observing creep under the picked stress at successively higher temperatures until an increment has been reached at which the specimen shows creep rates beyond those of interest.

In "step-up-temperature" tests, the cumulative effects of the stress and temperature upon the condition of the specimen at increments below the first one at which sensible creep occurs usually is negligible in its effect upon the resulting creep rate considering the use made of the results. The principal effect apt to be observed is a reduction of the amount of primary creep measurable at temperature increments beyond the first. The constant rate of creep, at temperatures where constant rate is attained, does not seem to be materially affected. Since the rate of creep tends to increase exponentially with temperature, it seems reasonable that the comparatively minute effects of creep at the lower temperature increments do not have much influence upon the behavior when a temperature high enough to produce sensible creep is attained. If evaluation of total creep (i.e., sum of primary and secondary creep for a given time) is desired, it is advisable to use a separate specimen for each temperature. When rates of the order of 0.1 per cent total plastic deformation must be taken into account, it becomes highly essential that the primary creep be given full consideration, since 0.1 per cent of creep often occurs during the early stages of a test under conditions yielding an almost nil rate during the secondary period.

The length of time necessary for the determination of a creep curve to be used for comparative purposes cannot be categorically stated. It not only depends upon the temperature and stress at which the test is made, but also upon numerous considerations involving the material of test, the purpose for which the data is to be used, and the progress of the creep curve as it is being observed. If 0.1 per cent plastic deformation is the tolerance limit and a primary creep of this amount occurs during the first two hundred hours of testing, there may be no point in continuing the test beyond such time, even though there is a tendency toward a slow secondary creep which might take

an additional thousand or more hours to evaluate reliably. The alternative situation also arises in which a material suspected of inclinations toward thermal instability, during the course of several thousand hours of testing might attain an exorbitant creep rate unforeshadowed by the early stages of testing. This tendency, fortunately, does not seem to be typical of many of the materials dealt with in creep testing and tests up to one thousand hours duration suffice for many purposes of evaluation.

Step-temperature tests at three temperatures generally fix the limiting creep temperature for a material, provided the temperatures are well chosen. The approximate creep character of most types of material has been well enough surveyed to enable us to do this. In the case of ferritic steels, the viscous transformation takes place over about a 200 degrees Fahr. (110 degrees Cent.) interval. At a load of 10,000 pounds per square inch, tests at 800 (425), 900 (480) and 1000 degrees Fahr. (540 degrees Cent.) usually give sufficient data for interpolating the limiting temperatures for 0.1 per cent and 1.0 per cent creep in 10,000 hours to within the limits of the accuracy of temperature measurement and control. The following table summarizes the schedule of loads and temperatures useful in determining the temperature creep relationship for certain types of materials.

Testing Temperatures

Testing Loads, p.s.i.:	30,000	10,000	3000
Ferritic Steels ..	700, 800, 900° F.	800, 900, 1000° F.	900, 1000, 1100° F.
Austenitic Steels.	1000, 1100, 1200° F.	1100, 1200, 1300° F.	1200, 1300, 1400° F.
Brass and Bronze	....	300, 400, 500° F.	400, 500, 600° F.

Certain variations of composition and treatment sometimes make it necessary or desirable to test at an additional temperature 100 degrees Fahr. (55 degrees Cent.) above or below the ranges shown in the table. This, however, is not often necessary since the character of most of the materials being tested is by now sufficiently well known to anticipate the three optimum testing temperatures. It should be noted that a 100 degrees Fahr. (55 degrees Cent.) shift in the testing temperature range corresponds to about a 3 to 1 shift in the testing stress. This corresponds approximately to the reduction of creep stress due to a 100 degrees Fahr. (55 degrees Cent.) increase in temperature. This ratio has been found to hold approximately for

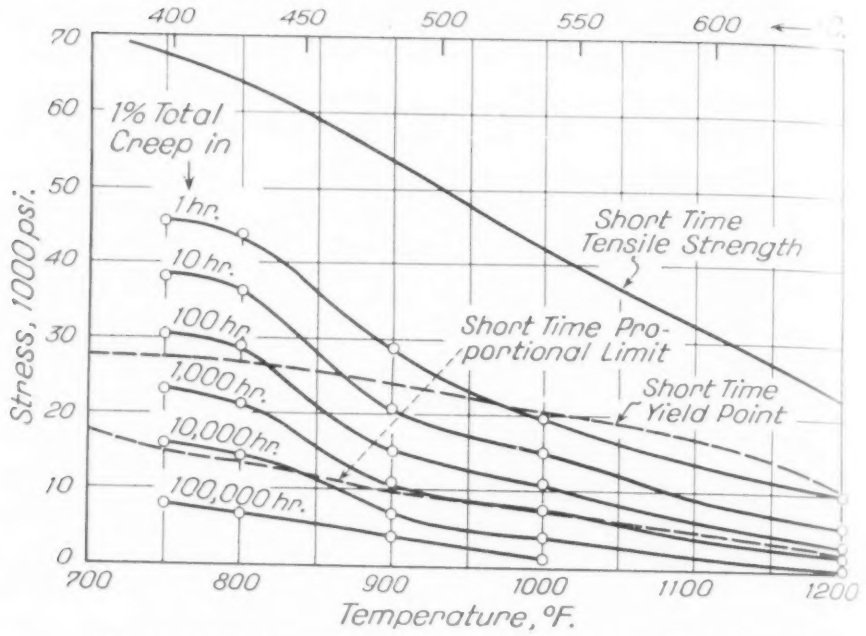

Fig. 14—Creep Stress Curves for Carbon Cast Steel With Short-Time Tensile Strength, Yield Point and Proportional Limit Curves Drawn in for Comparison. Kanter and Spring (59).

a number of different classes of materials in their "critical" creep ranges.

Working Stresses

The practical objective of creep testing and its interpretation is in the main to establish a rational basis for the selection of working stresses to be used with various materials. There are two distinct types of application for which creep information is desired. Where design is dictated by allowable stress, and the only consideration with regard to creep is that deformation does not approach the rupture point of the steel, comparatively coarse measurement suffices. The life of such high temperature equipment is generally expected to range from 2 to 5 years. The other type of application, where design is dictated by allowable distortion, requires stresses at which creep will not exceed 0.1 per cent during a life of 10 or 20 years. The permissible creep rates for various parts in steam plants have been listed by Baumann (22) as follows:

		Inches per Inch per Hour	Approximate Time for 1% Total Creep
(1)	Turbine disks pressed on shafts	10^{-9}	
(2)	Bolted flanges, turbine cylinders	10^{-8}	
(3)	Steam piping, welded joints, boiler tubes ..	10^{-7}	10 years
(4)	Super heater tubes	10^{-6}	1 year

The range of creep rates to be considered in design is thus a thousand fold. The range of this scale in relation to short-time and long-time tensile properties is illustrated by Fig. 14 representing data by Kanter and Spring (59) for carbon cast steel. This diagram illustrates the ratio in which stress is limited by permissible creep rate. Low stresses indeed are indicated when the application limits creep rate to 10^{-9} inches per inch per hour.

A number of investigators have attempted to relate the results of short-time tests with creep tests. In general, the short-time tensile strength and elastic limit are not reliable guides to small creep rates at temperatures where constant rates of creep may be observed. Up to the temperature where the rate of creep rapidly approaches zero, tensile strengths and elastic limits may be used as a basis for proportioning working stresses. Working stresses for carbon steels up to 650 degrees Fahr. (345 degrees Cent.) are determined by applying a factor of safety upon the atmospheric tensile strength. Above this temperature steady creep rates are determinable at stresses much less than short-time values and it is necessary to proportion working stresses according to creep values.

It has been suggested (39), (21), (47) that creep data for application involving creep stresses for rates of the magnitude 10^{-6} might be calculated from short duration tests varying from a few hours to several days. Considering the large variety of materials for which working stress data are required, the practicality of establishing rules for accomplishing this end is worthy of much study. This phase of creep test interpretation is still developing and as yet cannot be depended on as a reliable guide to the information that long duration tests of several thousand hours will develop.

The direct application of creep stresses as interpreted from laboratory tests of long duration to the design of equipment has not proven consistent with actual experience in many cases. As a result many engineers and metallurgists have become skeptical as to the practical value of creep testing in its relation to design problems. This situation has been brought about, no doubt, partly because attempts have been made to apply creep data with an insufficient understanding of the mechanics involved. When the background of experience which now dictates the choice of working stresses for high temperature is correlated with creep measurements, an applied mechanics of creep may then evolve and thoroughly rationalize the choice of working stresses. Standardization committees sponsored by the

A.S.M.E and A.S.T.M are attempting to set high temperature working stresses for various materials on a basis of creep data. This task is proving very difficult on account of great disagreement between creep values in the literature for supposedly similar material. Much of this discrepancy is due to nonuniformity of material and its treatment. Differences in testing apparatus and procedures employed by various investigators is accountable for spread of results. A tentative test code for tensile creep tests has been adopted by the American Society for Testing Materials (24), which, if strictly adhered to, should tend to produce better accord among the various creep investigators and eventually result in data thoroughly suitable for the critical application of creep toward which the designer is striving.

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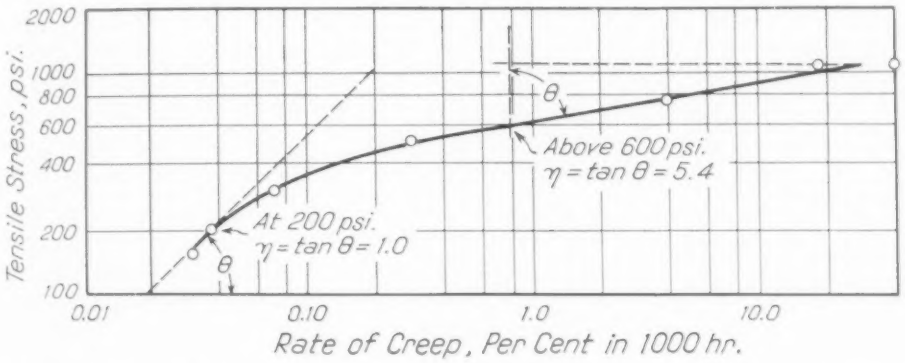
DISCUSSION

Written Discussion: By H. F. Moore and B. B. Betty, University of Illinois, Urbana, Illinois.

Mr. Kanter's critical discussion of Bailey's formulas for creep in terms of stress is extremely interesting. The writers of this discussion agree with Mr. Kanter's criticism of Bailey's fundamental equation

$$v = a\,\sigma^n$$

as given in equation [17] of Mr. Kanter's paper. As the authors of this discussion read Mr. Bailey's paper, he bases this equation on creep determinations under relatively high stress. There is submitted herewith a creep rate-stress diagram for tensile test of commercial lead, a diagram rather typical of diagrams obtained for several lead and lead alloys at the University of Illinois. This diagram is plotted on log-log paper. It will be noted that the value of



n is the slope of this curve at any pont, — that is, $n = \tan \Theta$ in the diagram. At high stresses for this particular experiment n was found to have the value of 5.4. At lower stresses its value approaches 1. That is, at lower stresses the equation for viscous flow represents the actual creep of the lead tension specimen fairly well. While the determination of precise formulas and the values of constants will be a matter of much experimentation, it seems to the writers of this discussion that Mr. Kanter's idea of quasi-viscous flow, which, at low stresses, becomes nearly a pure viscous flow is a better basis for developing creep formulas than those suggested by Bailey. Whether the pure viscous flow is a phenomenon of grain boundary flow, of flow along slip planes, or a combination of the two is, of course, a matter of further investigation.

Written Discussion: By F. H. Norton, Massachusetts Institute of Technology, Cambridge, Mass.

Mr. Kanter has given us in his paper not only an excellent review of the modern investigations on creep, but has digested the results and presented them in a most readable form. In this discussion any criticism is directed at the original papers cited rather than at Mr. Kanter's presentation of them.

In the first place, a general survey of the investigations cited would indicate that the essential difference between viscous and plastic flow is not clearly defined. As I understand it, viscous flow can only occur in a medium which acts homogeneously, such as oil or heated glass. Flow occurs even at the lowest

stresses because such a medium is incapable of supporting any shear. The flow is expressed by the equation:

$$V = kS$$

where V = the velocity of flow
and S = the stress

A great deal of work has been done on viscous flow in glass, and it is known that the time extension curve for a glass rod is strikingly similar to that for a piece of steel. There is the same initial portion of the curve where the flow rate decreases and then a portion where the rate is constant. At low rates of flow the initial period may continue as long as 500 to 1000 hours. Glass also shows a relaxation when the load is removed, exactly like a metal. Glass, however, is distinctly different from a metal when we observe the rate of flow under varying loads; for with glass, there is a true viscous flow, with the flow rate proportional to the stress, and this relation is followed exactly under a very wide range of flow rates. It used to be thought necessary to assume an actual change in the viscosity of glass with time to account for the change in rate of flow in the initial period, but due to some recent work carried out in our laboratory by Dr. G. J. Bair (*Journal* of the American Ceramic Society, Dec. 1936), it has been shown that this assumption is unnecessary, and that we may assume a constant viscosity. The change in rate can be accounted for by an elastic strain which is but sluggishly relieved.

Plastic flow occurs in any nonhomogeneous medium, as hard grains embedded in a softer matrix. Examples of this are clay and water mixtures, and metals above their minimum recrystallization temperature. In this type of flow it is generally considered necessary to have a certain value of stress to produce a permanent strain, but this is by no means a generally accepted fact, as our means of measurement may not be sufficiently good to determine the small strains actually produced under very small stresses. In general the plastic flow may be expressed by the following equation:

$$V = kS^n$$

In this equation the exponent n averages about 4 for flow rates reached in our ordinary testing, but it varies over quite a wide range as we change the temperature and stress; so that there is not a definite law of plastic flow but only an approximate one. Some interesting tests have recently been made in our laboratory on the flow of heated porcelains, as reported in the *Journal* of the American Ceramic Society, page 129, May 1936. The flow characteristics of the porcelain, which consist of crystals and a glassy matrix, are in every way similar to metals, showing the plastic type of flow, a strain hardening period, and finally a uniform rate. We are then evidently dealing with the same type of flow in materials as different as porcelain and steel, and while the internal structure may be entirely different in the two cases, the flow mechanism is the same. This is a fact of great importance.

If the flow rate of a steel is measured over a wide temperature range, we will find definite regimes of flow. The lowest will be an elastic condition where a given stress will produce a definite strain without continuous flow. As a temperature is reached which is generally supposed to be the minimum recrystal-

lization temperature, we enter the plastic regime, which continues up to the melting point of the metal. Above this temperature we have viscous flow or at high rates turbulent flow. A thorough study of the plastic flow is urgently needed to clear up many uncertain points. It is not necessary to make these tests on metals, as a wider range of conditions could be obtained by using, for example, a varied mixture of wax and fine sand. With pure wax there would be viscous flow, and as the amount of solid particles was increased there would be plastic flow. But whether or not there is a sharp transition between the two regimes, we have no way of knowing at present.

The work of B. Chalmers on single crystals of tin (*Proceedings of the Royal Society*, 156A, p. 427-443, Aug. 17, 1936) shows that at low rates of elongation the flow is viscous, but above a certain rate it becomes plastic. This fact has been observed by J. A. Fellows, (Massachusetts Institute of Technology, Master's Thesis) on a number of pure multicrystalline metals. This change in flow type can be shown clearly by plotting the log stress against the log rate of flow. A 45-degree slope indicates a viscous flow and a lower slope, plastic conditions. The sudden increase in slope at the low rates is clearly indicated. This fact has also been noticed by several experimenters on steel. This substantiates the conclusion of Mr. Kanter that we should not use the log-log plot for extrapolation. We are, however, justified in using it to interpolate, and for this purpose it is very convenient.

The initial portion of the creep curve is of particular interest and has been discussed quite thoroughly by Mr. Kanter. It seems to be generally accepted that the decrease in creep rate with time is due to a strain hardening phenomenon. But how can we explain that exactly the same thing happens in a piece of glass or porcelain, where the conception of strain hardening is not as well defined as with metals? A very interesting paper treating this subject has recently appeared in the Russian periodical "Metallurg" for January 1934, by Professor S. A. Oding, which was not cited by Mr. Kanter. In this paper equations have been worked out for the rate of hardening by strain and the rate of softening by the temperature, these being equated to give results which seem to compare fairly well with actual conditions. The equations are too long to reproduce in a discussion, but this work would seem to be one of the most successful mathematical attacks on the mechanism of creep.

Chalmer's paper also brings out the fact that at low flow rates, that is, viscous flow, the rate decreases exponentially with time, but at higher rates the flow rate remains constant after a certain period. This may explain some of the differences in results on long time creep tests, by different experimenters. Chevenard, and McVetty, for example, generally find a decreasing rate of flow over long periods, while others have shown a uniform flow rate. In this connection, it should be kept in mind that small changes in flow rate can be accounted for by a temperature change of only 4 or 5 degrees Fahr. In the long-time tests base metal couples tend to change in calibration, and often in such a direction as to allow the specimen temperature to decrease with time, thus producing a continuously decreasing flow-rate.

It would seem that our present data on creep, due to errors in testing and variations in the metals, is not sufficiently reliable to form a basis for a mathematical analysis of the shape of the strain-time curve or the stress-rate

curve. For example, if Figs. 5, 6 and 7 were plotted from another set of data, quite different conclusions might be arrived at.

Oral Discussion

A. NADAI:[1] I would like to express my appreciation for the paper that we have just heard, and to mention several points in this paper which probably deserve the attention of engineers who are utilizing the creep data. I wish to support Mr. Kanter with reference to the term viscosity which he has used in his recent paper and which has been found very useful where we have a connection between the speeds of deformation in machine parts and the stresses producing distortion in these parts. We have heard already today a few remarks about the so-called viscosity coefficient. It is my impression that this viscosity coefficient can clearly be defined at least for a given amount of strain hardening.

The phenomena of the creep of the metals at elevated temperatures are special cases of the plastic deformation of solids in which in general stress σ is dependent on two important variables, namely, the plastic strains ϵ (plastic unit elongation in a tensile test) and the plastic rates of strain $u = d\epsilon/dt$ (the velocity u with which a strain changes with the time t, strain rate or creep rate). We have therefore:

$$d\sigma = \frac{\delta\sigma}{\delta\epsilon} . d\epsilon + \frac{\delta\sigma}{\delta u} . du$$

The rate of change of stress with the plastic strain when the strain rate is not changed is:

$$\left(\frac{\delta\sigma}{\delta\epsilon} \right)_{u = \text{const}} = \Psi.$$

This quantity expresses the modulus of strain hardening. The rate of change of stress with strain rate at a given strain ϵ = constant is:

$$\left(\frac{\delta\sigma}{\delta u} \right)_{\epsilon = \text{const}} = \Phi.$$

This quantity defines the coefficient of viscosity of the metal. Both Ψ and Φ may in general be functions of ϵ and u. In many cases, however, simple expressions may be substituted for them.

With reference to creep, we thus have two important independent variables, strain ϵ and the rate of strain u; and I think the correct definition for such quantities as Ψ or Φ is probably to take the slopes of the surface which will connect these two variables with stress σ. I prefer to have this definition for the viscosity coefficient Φ which is different from the definition by Mr. Kanter who has introduced a value of Φ in his paper by dividing stress σ by the rate of flow u.

The other point which I wanted to mention is that it is very interesting to see that Mr. Kanter's Fig. 6 showed nearly a straight line when he plotted tensile load against minimum creep rate. The range of rates over which this was plotted is a comparatively small one, the creep rates covered a range be-

[1]Westinghouse Electric and Manufacturing Co., East Pittsburgh.

tween 10^{-3} and 10^{-5} so that the speeds have been changed not much more than one thousand times.

I may mention that in connection with some cooperative work the joint A.S.M.E.-A.S.T.M. Committee on the effects of high temperatures, we happened to have a chance to investigate a special carbon steel at 850 degrees Fahr. which has been carefully tested at Battelle Memorial Institute and in other laboratories in long time tests. Some of these latter tests are running more than two years, so that there was available valuable information obtained in long time tests on this material. In our Laboratories we prefer to include also rapid tests in our comparison of data. The points corresponding to the long time and to our rapid tests were plotted in a common curve for a constant strain. The rates of flow were plotted on a logarithmical scale. Stresses were plotted in a simple scale. The range in which test values could thus be compared was 10^8. Speeds were changed about one hundred million times. A good portion of the curve showing stress as a function of the logarithms of the strain rate was pretty straight but there was a very definite indication that the straight line would become a curve at small and at very large strain rates.

The conclusion which I want to bring out at this moment is that this special carbon steel at 850 degrees Fahr. showed in a certain range of speeds a behavior similar to that expressed in Mr. Kanter's Fig. 6; stress was a logarithmical function of the strain rate at a given constant strain. For higher and lower rates the straight line becomes a curve.

W. A. MADDOX:[2] It would be very interesting to know if we may extrapolate from creep curves such as these in order to predict just how much stress relief does result when we stress relieve in a furnace some structure such as a casting or welded detail. Data is sadly needed that is applicable to cast iron, cast steel and welded structures.

In short, since the stresses we wish to relieve may range upward to 45,000 pounds per square inch, may we by heating at 1200 degrees Fahr. for one hour relieve our stresses to or below 10,000 pounds per square inch, a reasonable rate of heating and cooling being assumed?

Encouragement for making such a deduction is at least indicated by the one hour creep curve, Fig. 14.

Author's Reply to Discussion

The discussion which the present paper has elicited is most generous and conveys the feeling that engineers today stand upon much common ground in their understanding of the creep problem. With the vast fund of experimental data which has made its way into the literature, and systematic scrutiny to which this information is being subjected, a complete rationalization of the problem seems not far away.

The work on lead and lead alloys in which Moore and Betty are engaged promises to serve the purpose of establishing certain of the essentials in the fundamentals of creep. Their curve for rate of creep versus tensile stress, offered in this discussion, well justifies the conception of viscous creep. Their data when plotted according to the scheme of Fig. 7 shows the lower points

[2]Cincinnati Milling Machine and Cincinnati Grinders, Inc., Cincinnati.

becoming parallel to the stress ordinate indicating most clearly the behavior of pure viscosity at small stresses.

Mr. Norton's discussion gives us a concise statement of the differentiation between plastic and viscous deformation. In adopting the term "viscous" with reference to creep in metal, perhaps it should be borne in mind that we are merely drawing an analogy to the behavior observed in the deformation of homogeneous media. For this specific reason the term "quasi viscous" has been employed. The author is well in agreement with Mr. Norton that the strain-time curve and the stress rate curve have merely an empirical significance and any conclusions drawn from these diagrams are to be regarded as such. Figs. 5, 6 and 7 are merely shown as typical illustrations of what is found when plotting the best available data for a comparatively stable steel.

The proper definition of the viscosity coefficent for metals, as Dr. Nadai points out, requires it to be a derivative of the stress with respect to rate of deformation at a constant value of strain hardening. The coefficient introduced in connection with equation [26] is not strictly in accord with the precise definition, but has the dimensions of viscosity and is applied in correlating the so-called viscous creep associated with constant creep rates. A coefficient as defined for equation [26] affords an expedient means of correlating stress with creep rate for extrapolation purposes.

Mr. Maddox raises the question of using creep data for predicting stress relief in cast and welded structure. Qualitatively, creep curves such as shown in Figs. 13 and 14 would seem useful in predicting the temperature or range of temperature where such relief can be produced in reasonable time. The exact calculation of the time for relief exorbitant stress to a tolerable value is essentially a relaxation problem and should be treated through the use of relaxation data.

DAMPING CAPACITY—ITS VARIATION AND RELATION TO OTHER PHYSICAL PROPERTIES

By G. R. Brophy and E. R. Parker

Abstract

Damping capacity (plasticity) of steel is greatly influenced by such factors as stress, chemical composition, structure, heat treatment, mechanical work, and temperature. These factors are dealt with briefly, and in addition the relation of damping capacity to fatigue and creep performance is shown.

Damping capacity is particularly important in determining the differences in behavior of steel, when repeatedly stressed, to the effects of overstress, stress concentration and speed.

The order of creep resistance of a series of steels at elevated temperatures may be predicted by determining the damping capacities at the desired temperatures, provided there are no structural changes other than grain size.

INFORMATION on damping capacity has been in use by designing engineers for considerable time, and its importance in connection with vibration phenomena has been discussed quite completely in a number of publications. (1), (2), (3), (4).[1] As a physical property controlling the performance of metal, and especially its orderly variation with changing conditions, it is perhaps less familiar. The property is attributed to plastic action which can be measured even within the so-called elastic range of metals down to the lowest stresses.

It appears, therefore, that within our experience reaction to stress is both elastic and plastic deformation. At stresses below the elastic limit the plastic deformation is but a very small portion of the total, but increases rapidly to predominant amounts above.

Damping capacity is defined as the "energy dissipated as heat by a unit volume of metal during a completely reversed cycle of unit

[1]The figures appearing in parentheses pertain to the references appended to this paper.

A paper presented as part of the Symposium on the Plastic Working of Metals, Eighteenth Annual Convention of the Society, October 19 to 23, 1936. The authors are members of the staff of the Research Laboratories, General Electric Co., Schenectady, N. Y.

stress," which is the energy lost in the change from potential to kinetic energy. It is manifested in various ways as, for instance, when a thin strip of metal is set in free vibration there is a continuous decrease of amplitude with time. This property has been variously termed by a number of authors, the more common terms being internal friction and dynamic ductility.

Since damping is a measure of plasticity it should be possible to correlate it, at least qualitatively, with the other properties which

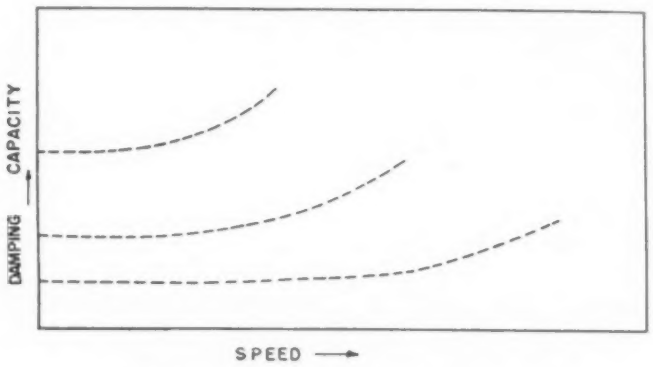


Fig. 1—Possible Influence of Frequency on Damping Capacity.

depend upon plasticity, and since it can be quickly and accurately determined, more information on plasticity and its change may be obtained by the study of damping than by the examination of other properties.

In any metal damping capacity may vary greatly depending upon such factors as stress, chemical composition, structure, heat treatment, mechanical work and especially temperature. Combinations of these factors, which are usually encountered, complicate the interpretation of damping capacity determinations, until the influence of each is understood.

This paper deals with the influence of these various factors and, in addition, covers the relation of damping capacity to endurance properties and creep resistance of steel.

Influence of Frequency

The frequency of stress alternation has no effect upon the energy lost provided the temperature of the test piece during the test remains constant. Since the energy lost is through friction resulting in heat, it is entirely possible to obtain a frequency sufficiently

great that the system cannot dissipate the heat as rapidly as generated so the temperature of the bar will rise. If so, the damping capacity will rise. A material of high energy loss will probably show an earlier increase in temperature with frequency than one of low damping capacity. Fig. 1 illustrates the probable trend. The real significance of this is realized in fatigue testing at high speeds which will be treated at length in a later part of this paper.

Variation with Stress

Characteristic specific damping capacity curves show a continuously decreasing energy loss with decreasing stress which, however, may retain a finite value down to zero stress. There are some exceptions to this as illustrated in Fig. 2, which shows a damping

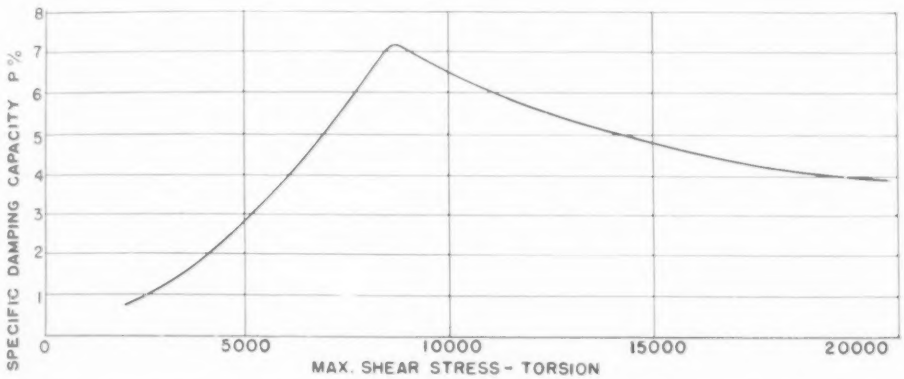


Fig. 2—Damping Capacity of a 5.2 Per Cent Chromium—0.5 Per Cent Molybdenum Steel Air Cooled from 900 Degrees Cent.—650 Degree Draw.

curve peculiar to chromium steels containing over 3 to 4 per cent of chromium. No ready explanation is obtained for the increasing loss as the stress falls through a certain range.

Composition

Chevenard (5) states that pure metals have high damping capacities and that as the number of alloy additions increases the damping capacity decreases, and that the first additions of an element are more effective than the later ones.

Heat Treatment

For a given composition the annealed alloy has a higher damping capacity than one which is hardened by heat treatment. Thus,

with steel the hardest condition obtainable by quenching is most nearly elastic and shows the least energy loss, whereas reheating for softening increases the loss.

This relation at room temperature, however, is modified by grain size. Increasing the grain size of steel will result in raising the damping capacity. This is more particularly noticeable in steels which are quenched and tempered than in those which are annealed. Fig. 3.

It has been noticed, too, in some cases that an annealed steel subcritically quenched shows a marked reduction in damping capacity

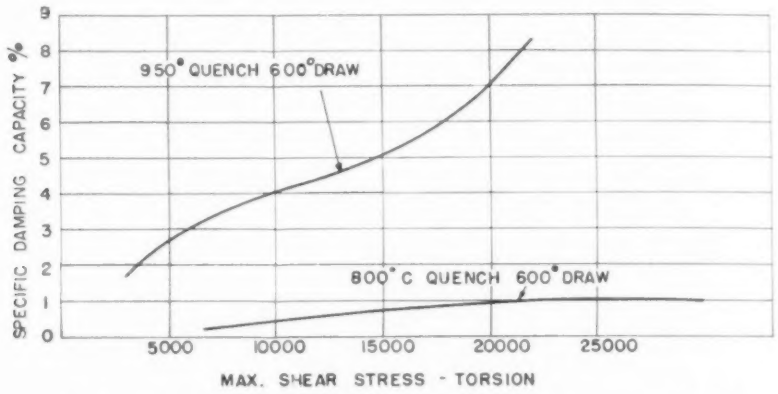


Fig. 3—Influence of Grain Size on the Specific Damping Capacity at Room Temperature of a 3.1 Per Cent Nickel, 1.28 Per Cent Chromium, 0.35 Per Cent Carbon Steel.

indicating a hardening. This response is especially strong in temper brittle steels, as would be expected, and may be significant in the performance of steels under repeated stress where alternations are rapid.

Temperature

Chevenard (5) states that "for pure metals and normal alloys .....the energy loss increases with rising temperature and obeys a simple law." As a general statement, perhaps, this is true, but for steels this increase at a given stress is not continuous as shown, for example, in Fig. 4, where damping curves are given for an annealed 0.36 per cent carbon steel tested at different temperatures both on heating and cooling.

By cross plotting at a given stress a damping temperature curve is obtained and such a plot is shown in Fig. 5 at 6,000 pounds per square inch (2). It is seen that upon heating the damping capacity at first rises, but at approximately 300 degrees Cent. there is a sudden

drop which is followed by a rapid rise. Upon cooling the damping capacity at any temperature is lower and but a slight rise is noted at 300 degrees Cent.

It is believed that some soluble constituent is dissolved at 300 degrees Cent. upon heating which lowers the damping suddenly, and upon cooling is precipitated more or less completely, depending upon the cooling rate and in a finely divided state. This is confirmed

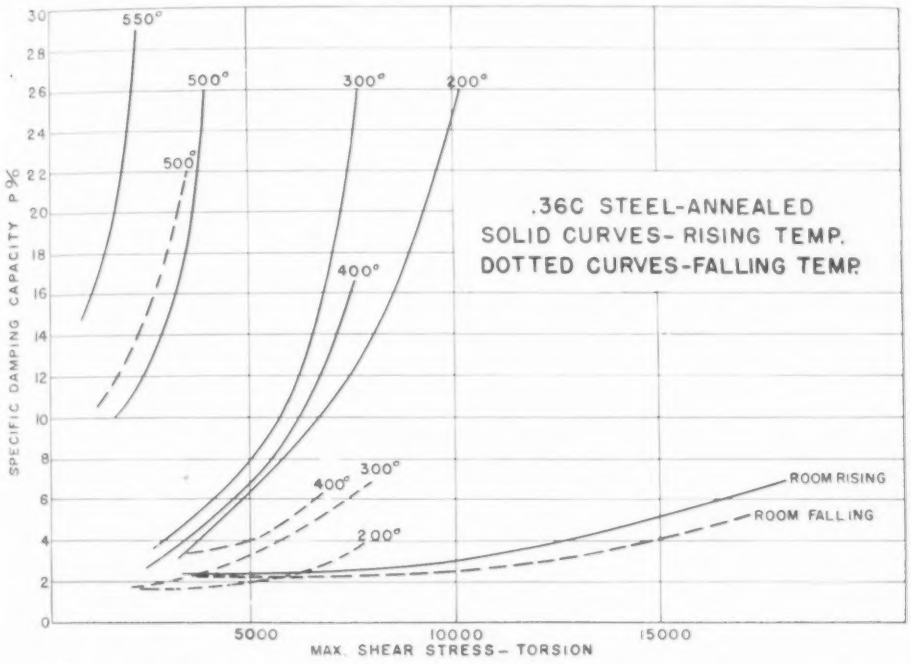


Fig. 4—Damping Capacity Curves for a 0.36 Per Cent Carbon Steel Determined at Different Temperatures Upon Heating and Cooling.

by the fact that after water quenching from 600 degrees Cent. both heating and cooling curves show no reversals and the steel has a much lower damping capacity over the full temperature range. Reheating to 600 degrees Cent. and cooling slowly in the furnace restores the initial characteristics.

Upon repeated heating and cooling of the annealed bar the heating curves tend to merge with the cooling curves. The reversal on cooling is eliminated entirely, and, on heating, occurs at lower temperatures, although it is believed that it will never be eliminated, but will approach a limiting temperature of approximately 200 degrees Cent.

Grain size again modifies the plasticity of steel at elevated temperatures, but in the opposite way to that found at room tem-

perature which was shown in Fig. 3. Above a certain temperature, which varies with the steel composition, it is found that increasing the grain size decreases plasticity as measured by damping capacity.

Fig. 6 shows the damping curves at room temperature and at

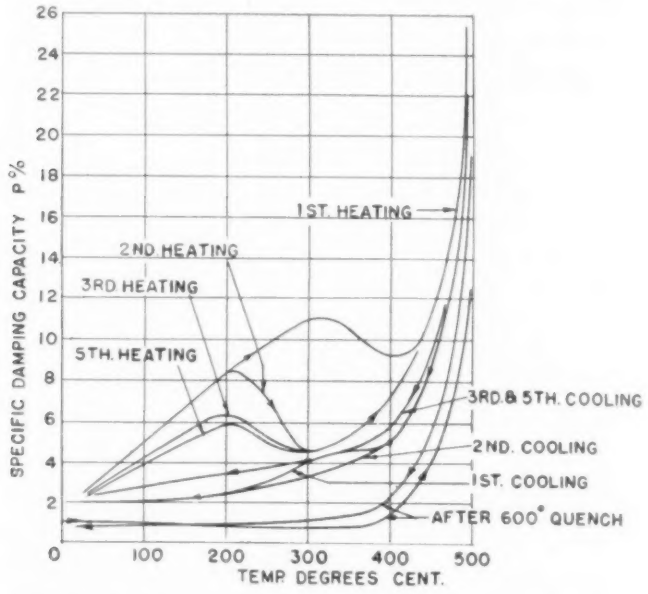


Fig. 5—Damping Capacity at 6000 Pounds Per Square Inch Determined at Different Temperatures on Rising and Falling Temperatures.

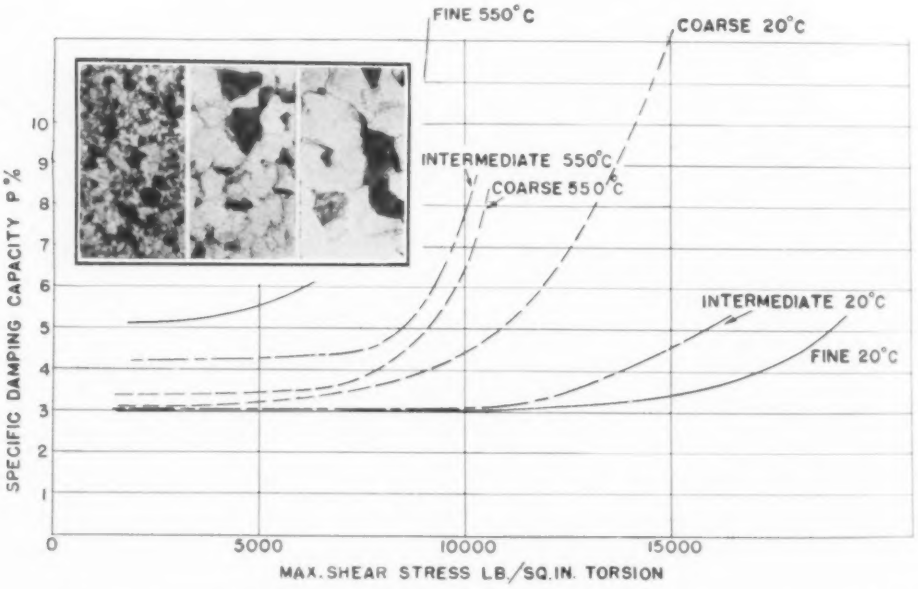


Fig. 6—Influence of Grain Size on Damping Capacity at 20 Degrees Cent. and 550 Degrees Cent.

550 degrees Cent. for an alloy steel of the following composition:

Carbon 0.20 Per Cent Chromium 1.75 Per Cent Tungsten 1.75 Per Cent

which was slowly cooled from three different temperatures to develop different grain sizes as shown in the micrographs.

Correlation with Other Physical Properties

Attempts have been made to correlate damping capacity with tensile strength, hardness, notched impact resistance, fatigue and creep resistance. No relation to impact resistance has been found, and no general relation to tensile strength, hardness, or endurance limit, except that in any one steel, as these properties increase, damping capacity is lower (6), (7).

Fatigue

In previous publications (4) it has been shown that high damping capacity is a great protection to steel when subjected to alternations of stress in the presence of high local stress concentrations, but damage resulting from occasional stresses above the endurance limit, however, is greater (7).

As previously stated, the damping capacity probably would increase with speed of alternations of stress if the speed and stress were sufficient to raise the temperature of the test piece. This, of course, actually occurs during fatigue testing where the stresses are high. The bar heats sensibly and at high speeds may even show redness, in which case the damping capacity rises; and because it rises more heat is produced. This continues accumulatively to the failure of the bar. It has been noticed that some steels heat more rapidly and at lower speeds and stresses than others, and it has been determined that these have high damping capacity.

The apparent discrepancies existing in the literature upon the effect of speed on endurance limits can be explained, it is believed, by the difference in damping capacities of the steels tested.

For instance, Moore and Kommers (4) report no change in endurance limit up to 5,000 revolutions per minute, while Krause (9) recently reported but a slight rise up to 30,000 revolutions per minute. Jenkins (10), on the other hand, working with vibrating reeds, found endurance limits rising to a maximum at speeds varying from 10,000

to 20,000 reversals of stress per minute depending upon the steel and its heat treatment, and lower endurance limits beyond the maximum.

From our own tests (tabulated below) on a limited number of steels tested at 2,000 and at 20,000 revolutions per minute results similar to Jenkins' have been obtained:

Steel	Endurance Limit at 2,000 r.p.m.	Endurance Limit at 20,000 r.p.m.	Per Cent Loss or Gain at 20,000 r.p.m.	Per Cent Damping
0.37% C (coarse grain) quenched and drawn	45,500	41,500	−8.8	8
0.36% C (fine grain) quenched and drawn	42,500	42,500	0	5
3.0% Ni; 0.75% Cr; .32% C quenched and drawn				
A—quenched from draw	66,700	76,000	+13.9	3
B—slow cool from draw	63,000	72,000	+14.2	2.5

These results show that high damping steels have lower endurance limits and that low damping steels have higher endurance limits at high speeds than at low speeds. It may be that any steel

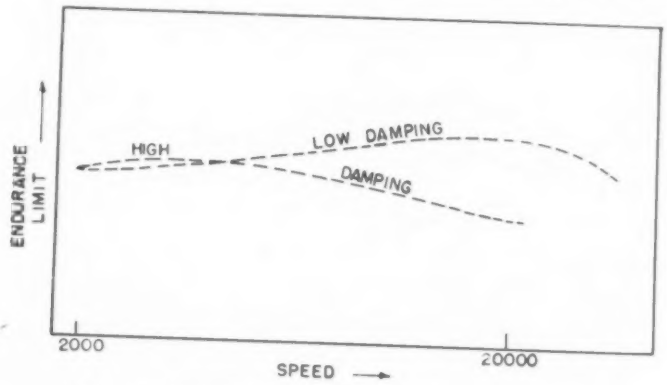


Fig. 7—Showing Probable Variation of Endurance Limits with Speed for Steel of High and Low Damping Capacity.

rises to a maximum value for endurance limit at some speed according to Jenkins' inference, and, if so, Fig. 7 illustrates the probable situation. The rise to a maximum is the result of heating to the "blue brittle temperature" at which the steel is stronger. A high damping steel would reach this temperature earlier than a low damping steel because of the difference in the amount of heat generated at any speed and stress. It was previously mentioned, however, that the change of damping capacity with temperature may be reduced in some steels by subcritical quenching as shown in Fig. 5 and as a result better performance obtained at increased speeds.

Notch sensitivity is increased by speed, but to a greater extent in the more notch sensitive materials, which have been shown to have low damping capacities (4), (7). The following data taken from tests illustrate this point:

Damping Capacity	Stress Concentration Factor at 2,000 r.p.m.	at 20,000 r.p.m.	Per Cent Increase Due to Speed
High	1.64	1.70	3.6
Medium	2.68	3.53	31.5
Low	3.32	5.54	64

Creep

As stated in the beginning of the paper, the continuous decrease of amplitude (damping capacity) of a freely vibrating bar of steel is the result of plastic deformation. Plasticity is also manifested by the continuous elongation of a bar of steel under constant load.

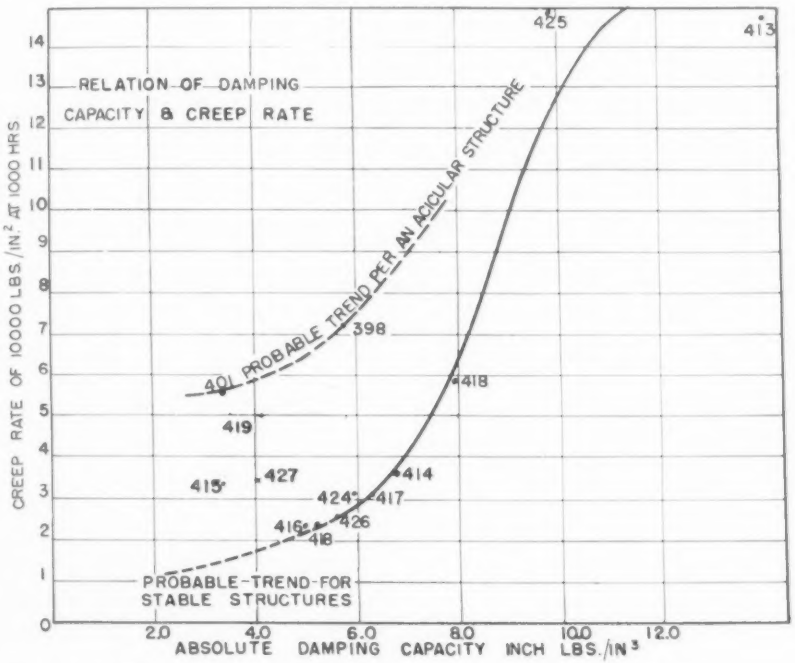


Fig. 8—Relation of Damping Capacity and Creep Rate.

It is logical to expect, therefore, at least a qualitative relation between damping capacity tests and creep tests conducted at the same temperatures. The relation is qualitative because the damping test is completed in a few minutes, whereas the creep test occupies thousands of hours.

Comparative tests, details of which will be the subject of a

future paper, have been made at 550 degrees Cent. on three alloy steels, each having been annealed at four different temperatures with the intention of developing a range of grain sizes. All bars were cooled slowly in the furnace from the annealing temperatures so that the cooling rates were approximately the same.

The structures developed are shown in Fig. 10, arranged according to the curves of Figs. 8 and 9. It is rather important to

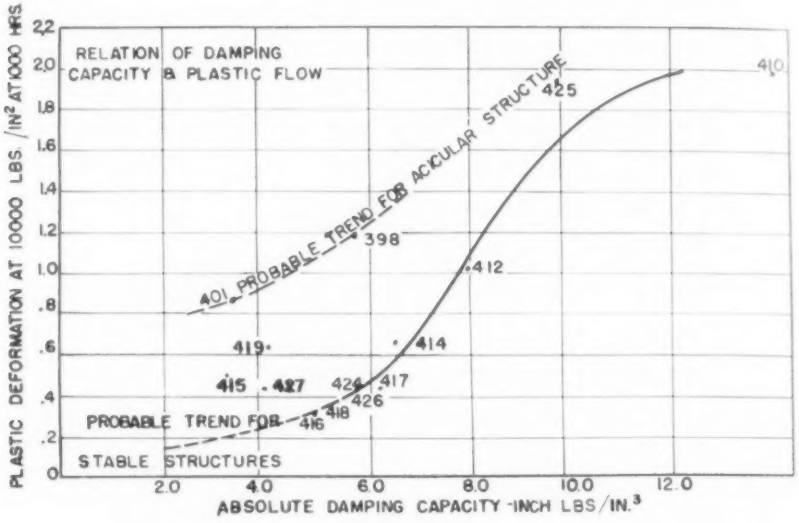


Fig. 9—Relation of Damping Capacity and Plastic Flow.

note that the very large grained samples have developed an acicular structure which we have termed unstable (at the test temperature of 550 degrees Cent.).

Items 401 and 398 are the same steel as items 416 to 419 except that they have been air cooled which results in a greater degree of acicularity, or instability.

Figs. 8 and 9 show the correlation obtained between damping capacity and creep rates, and damping capacity and the total plastic deformations respectively at the end of 1000 hours. The curves show the relation expected if both damping and creep are measures of plasticity; i.e., with decreasing damping capacity there is a decrease of creep rate and plastic deformation.

Points falling above the curve at low values of damping capacity indicate some change occurring at a critical grain size for the rate of cooling used. Reference to Fig. 10 shows this to be the result of a change from fine pearlite to an acicular structure. It is probable that much reduced rates of cooling for the grain sizes beyond the

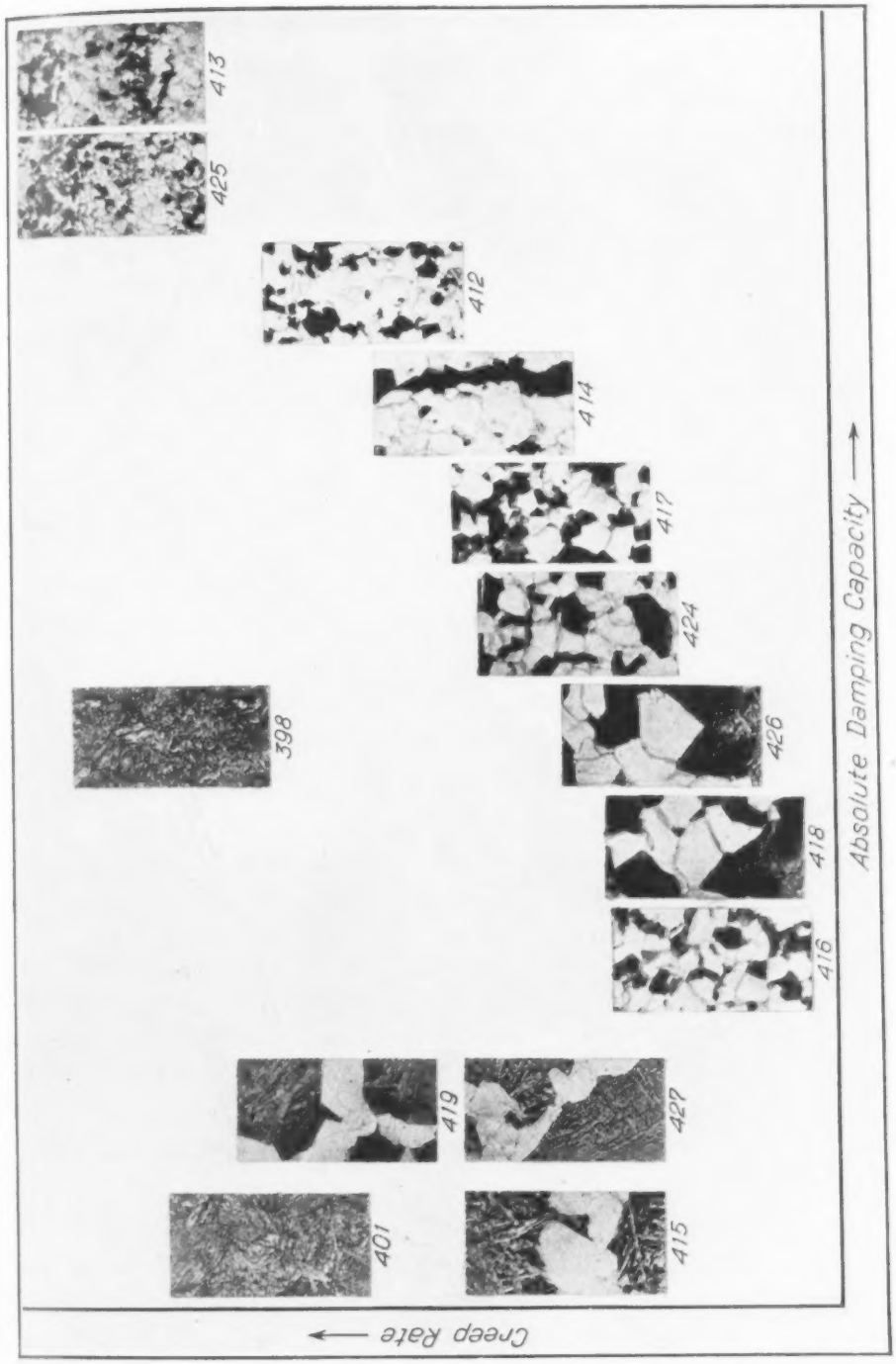


Fig. 10—Structures of Steels Tested for Creep and Damping Correlation, Arranged According to Curves of Figs. 8 and 9.

critical would result in fine pearlitic structures and the points would then fall somewhat as indicated by the dotted extension of the curve.

As a matter of fact the structure of the large grained samples may be considered to be the result of a mild quench and to belong, not to the stable curve, but to curves lying above. This is supported by the fact that the air cooled series fall on a separate curve higher than the others.

In general, the smallest grain sizes are weaker than the coarsest, but this relation does not hold except in the case of any one steel. In other words, no general relation is found to exist between creep and grain size unless composition is considered.

On the basis of damping capacity, on the other hand, a good relation is obtained provided the nature of the structure does not change in the series and, within limits, is independent of composition.

The temperature at which these tests were conducted was above the so-called equi-cohesive temperature for the materials studied, so that the general grain size-creep relation is valid. At temperatures lower than the equi-cohesive temperature the grain size-creep relation is reversed as is well known (11), but the damping-creep relation shown here in all probability still holds since both are measures of plasticity.

Conclusion

The study of damping capacity as a property of metals is, to say the least, extremely interesting and offers a means of understanding better the action of metals under stress. It is not proposed here that the damping test be used as a substitute for others, but in conjunction with them, and possibly as a means of surveying rapidly a large number of materials and thereby eliminating long and tedious tests on the unpromising.

As a property, it is of greatest importance in the control of the deflection and whipping of rotating shafts, and the surging of springs. Its variation with temperature appears to be a controlling factor in the fatigue and creep resistance of steel. Other relations will undoubtedly be found as there remains a number of problems to be investigated; for instance, a study of the relation of damping capacity to work hardenability, impact resistance, the relief of internal stress and others.

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ELASTIC PROPERTIES AND THEIR RELATIONSHIP TO STRENGTH AND TO STRAIN HARDENING

By M. F. Sayre

Abstract

Experiments recently carried on in this country and in Germany have shown that adjacent grains in a bar of iron, depending upon the angle between stress direction and crystal axes, will vary in effective modulus of elasticity from as low as 19 million to as high as 41 million, and that even larger variations exist in copper and brass. It is indicated in this paper that these stiffness variations must imply corresponding stress variations, and so may be of major importance in controlling the strength of the material. Cold work, for example, tends to develop a type of preferred orientation which corresponds to more favorable stress distribution relationships than with random orientation, and so gives a directional increase of strength. It is also suggested that these stiffness variations from grain to grain may aid in accounting for occasional cases of poor surface in cold pressing operations, and from another point of view, may help in explaining deviations between theoretical and experimental results in cases of localized stress concentrations.

SO MUCH work has already been done upon the mechanism by which strain hardening occurs in the cold working of a metal that it is difficult in any statement on the matter to avoid considerable repetition. It is believed, however, that new light has been shed upon this problem by certain recent experiments upon the elastic properties of metal single crystals. It has generally been assumed in the past that at least in metals with cubic atomic arrangement, the individual crystals could be considered elastically isotropic, with moduli of elasticity the same in all directions. Work done within the last few years by several investigators has shown that this is far from true, the moduli of elasticity in tension and in shear varying by as much as 3 or 3.5 to one, depending on the direction in which the stress is applied with respect to the crystal axes. Directional variations in stiffness of this magnitude will obviously have a very large influence upon the uniformity of stress distribution

A paper presented as part of the Symposium on the Plastic Working of Metals, Eighteenth Annual Convention of the Society, Cleveland, October 19 to 23, 1936. The author, M. F. Sayre, is associate professor of applied mechanics, Union College, Schenectady, N. Y.

within a material, and so upon the effective strength of the material. Those grains so oriented as to have a high modulus will be under severe stress, with lower stresses on adjacent, differently oriented grains. At individual points of contact between highly stressed grains, still higher stress concentrations will occur.

Apparently an initial effect of cold work will be to cause plastic yield at these points of high stress concentration, so causing stress equalization and increasing the effective strength for future loads in the same direction. For loads in other directions, an entirely different set of contact points come into play, so that the improvement in strength due to this will be in some measure directional. Continuation of the cold work develops a preferred crystal orientation, which on the basis of these moduli of elasticity variations should result in further equalization of stress between grains, and so increase the effective strength in favored grain directions. Meanwhile the cold working will also result in grain refinement, which should give a non-directional increase in strength.

This directional stiffness effect is not offered as a substitute for other factors in the mechanism of strain hardening. It should be treated simply as one additional factor which needs consideration. Its place in the general picture may be better seen if a brief review is given of these other factors. This may possibly best be done by considering first those factors which control the strength of a single crystal, and then going on to the more complex case of polycrystalline material.

Single Crystal Behavior

The inherent resistance to slip in pure metal along any individual crystal plane, assuming perfectly regular arrangement of atoms, is usually extremely small, and may approach zero. The atoms along both faces of the slip plane are all identically alike, and it is a matter of indifference to any one atom whether it mates up with the particular atom which is opposite to it across the plane, or with some other atom, now one hundred atoms distant. Atoms at the edge of the slip plane have more at stake, and as a result in this hypothetical simple crystal, surface tension along the edges of the slip plane may be a factor in resisting slip at least comparable in magnitude to the sliding resistance along the planes themselves.

Presence of even an extremely minute percentage of foreign atoms in solid solution, or of foreign crystallites, results in a surprisingly large increase in resistance to shearing slip. These cause

irregularities in the crystal network, and, according to the slip interference theory, by a simple mechanical process key the adjacent crystal layers together against movement. High strength of this type will be the result of chemical composition or of heat treatment, not of strain hardening.

A second, altogether different form of strengthening may, however, be accomplished by bending or twisting the single crystal, so warping the atomic planes. This again mechanically increases the resistance to slip. Sliding of one face with respect to an abutting face along a warped surface is evidently possible only if the slip along the principal surface is accompanied by many separate slips along intersecting surfaces, which will allow the two abutting faces to change in shape as they slide past each other along the warped surface. It is also likely to be accompanied by the development of sub-microscopic cracks and voids. The total amount of internal work for a given amount of elongation is therefore much greater than with the original crystal.

So far, reference has been made primarily to the forces required to produce slip along given crystal planes. As differentiated from this, the strength of the crystal grain itself in resisting elongation is dependent upon the angle between the slip plane and the direction of the force upon the bar. The intensity of shearing stress is maximum for an angle of 45 degrees, and this is usually regarded as the angle for minimum tensile strength. There seems, however, to be some experimental evidence tending to indicate that the inherent resistance to slip along crystal planes is increased by the presence of compressive forces perpendicular to the planes, and decreased by the presence of tensile forces, somewhat analogous to the conditions of solid friction. If this is the case, the angle of minimum strength will be greater than 45 degrees for compressive forces, and less than 45 degrees for tensile forces. For single crystals with hexagonal atomic arrangement and only one preferred group of slip planes (as in zinc) this relationship between the tensile strength of the bar and the angle between the applied force and the slip plane direction is experimentally very evident. For metals with cubic atomic arrangement, slip may occur on any one of several different families of planes, and that family of planes which is inherently weakest and also most disadvantageously placed will control. The directional variations in strength will be smaller than with hexagonal metals, but still present.

Behavior in Polycrystalline Material Under Stress

Polycrystalline metallic masses are, of course, of far greater commercial importance than these single crystals. Each grain in these polycrystalline masses however constitutes a separate single crystal, and except for the limitations imposed by the presence of surrounding grains, behaves as already described.

The problem, however, even for a purely homogeneous material, is much more complex than for the single crystal. One important difference comes from the fact that individual grains differ in their rate of elastic distortion under load, depending upon crystal orientation with respect to the direction of stress. Instead of being uniform over the entire cross-section, as was assumed for the single crystal, the stress distribution across the material is therefore necessarily markedly nonuniform from grain to grain. Furthermore, once plastic flow starts, sharp modifications occur in the initial stress distribution, fortunately usually in the direction of improved conditions.

The directional variations in modulus of elasticity in metal crystals upon which this nonuniform stress distribution depends, have been under active investigation by physicists for the past eight years. Masima and Sachs (1)[1] in 1928 reported upon the elastic properties of brass crystals of 72 per cent copper content. Goens and Schmid (2) have made similar studies of Armco iron crystals. Weerts (3) has reported on copper, Hanson (4) and Bridgman (5) on zinc, and Boas (6) on tin, among other investigators. Bruggeman (7) has summarized the work of earlier investigators, and then taken up the task of interpreting the relationship between the moduli of elasticity in different directions for single crystals, and the corresponding moduli of elasticity in different directions as experimentally determined in drawn wire, cold-rolled sheet, or other cases where preferred orientation is evident. A feature of Bruggeman's paper is an extensive bibliography of over sixty titles bearing on this general subject of moduli of elasticity variations with texture, or after various types of overstress.

The general form of variation of modulus of elasticity with orientation as found by Goens and Schmid for Armco iron may be taken as typical of other metals with cubic crystal lattice as well. Minimum values of Young's modulus (E) occur for stresses acting parallel to the three major crystal axes. For single crystals of

[1]The figures appearing in parentheses refer to the bibliography appended to this paper.

Armco iron so placed that the load acts parallel to any one of these three major axes of the crystal Goens and Schmid found Young's modulus to be 19,200,000 pounds per square inch. As the direction of the load is tilted away from these axes, the modulus increases, until a maximum value of 41,200,000 is obtained when the load acts along a space diagonal between the three axes. Fig. 1, taken from Goens and Schmid's paper, is a photograph of a three-dimensional model intended to portray this variation in modulus with direction, the distance in any direction between the center of the model and the

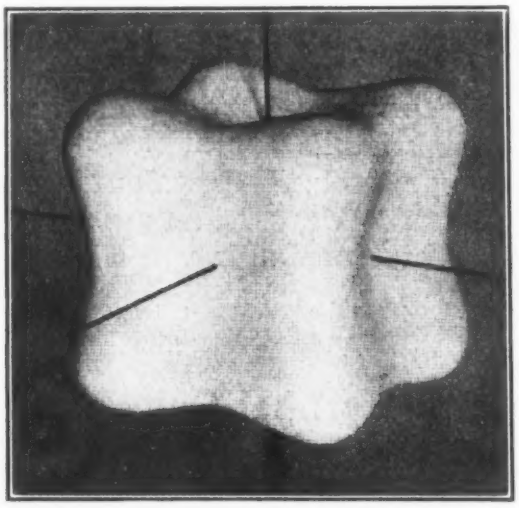

Fig. 1—Young's Modulus Diagram for Iron Crystals.

surface being proportional to the modulus of elasticity for loads acting parallel to this direction.

Modulus of elasticity in shear shows a reverse variation. For torques acting about any one of the three major axes, a single crystal will have maximum stiffness, the shearing modulus of elasticity being 16,800,000. As the direction of axis of torque shifts away from these positions, the stiffness decreases and the shearing modulus with it, down to a minimum value of 8,680,000 for torques acting about the space diagonal between the cubic axes. This variation is shown graphically in Fig. 2.

Table I, taken from Weerts' paper (3), but with values converted into English units, gives corresponding figures for a number of other common metals.

Only a partial degree of parallel orientation can be expected

Table I
Young's Modulus and Shear Modulus of Cubic Metal and Alloy Crystals at Room Temperature.

Material	Structure	Elasticity Modulus, E in 10^6 {111}	{100}	Shear Modulus, G in 10^6 {111}	{100}	Elastic Anisotropy E_{111}/E_{100}	G_{100}/G_{111}	Observer
Aluminum	Face-centered	10.81	9.24	3.55	4.12	1.18	1.13	Goens (1929)
Aluminum, 2 atomic % copper, heat treated	Face-centered	11.38	9.67	3.55	3.84	1.18	1.12	Karnop-Sachs (1929)
Gold	Face-centered	16.21	5.97	2.56	5.83	2.7	2.2	Goens (1929)
Copper	Face-centered	27.59	9.67	4.41	10.95	2.86	2.47	Goens-Weerts (1932)
Copper with 25 atomic % Gold	Face-centered	24.60	8.82	3.98	9.39	2.79	2.37	Sachs-Weerts (1931) Goens-Stenzel-Weerts (1933)
Copper with 28 atomic % Zinc	Face-centered	26.17	7.54	3.55	10.38	3.5	3.0	Masima-Sachs (1928)
Iron	Body-centered	41.24	19.20	8.68	16.78	2.15	1.94	Goens-Schmid (1931)
Tungsten	Body-centered	56.89	56.89	22.04	22.04	1.00	1.00	Bridgman (1925) Wright (1930)

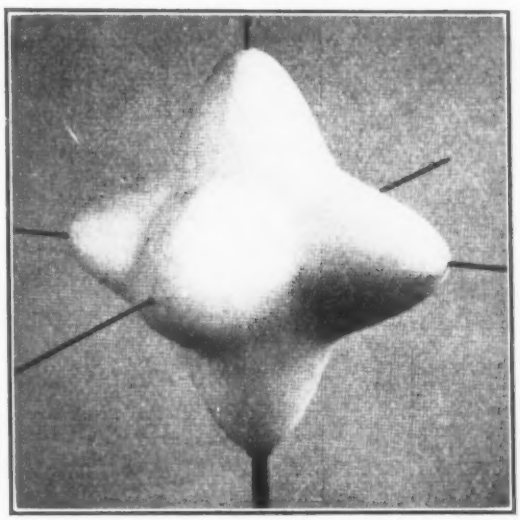

Fig. 2—Torsional Modulus Diagram for Iron Crystals.

in cold-rolled rod or sheet, even after very large amounts of cold work. Directional variations in moduli of elasticity as experimentally determined will therefore naturally be smaller than these extreme ranges of two or three to one found in the single crystals. Experimental variations with general order of magnitude ranging from 6

to 5 to 3 to 2, however, are given by Goens and Schmid (2) and by Bruggeman (7), and these correlate moderately well with values computed on the basis of X-ray determinations of the type of preferred orientation.

On cold drawn wire both in the straight form and as coiled into helical springs, equivalent evidence of presence of elastic anisotropy has been found in experimental work carried on at Union College. Here, due to the obvious difficulty of measuring the modulus of elasticity at a 45-degree angle to the length of the wire, which is the angle at which maximum variation could be expected, indirect methods have been used. For an elastically isotropic material, the relationships between Young's modulus E, the shearing modulus G, and the bulk modulus K are given by the equation:

$$E = \frac{9\,G\,K}{G + 3K}$$

This equation, and an equivalent one connecting Young's modulus, the shearing modulus, and Poisson's ratio, are often used as if generally valid. Actually, they hold true only for metals with random orientation. Furthermore, the ratios between E and G allowed by this equation are relatively limited in range. The bulk modulus is usually in the neighborhood of 0.80 E, whence G = 0.387 E. For K = 0.75 E, this relationship changes only to G = 0.391 E. For anisotropic material, however, a very wide range of G/E ratios is possible. For an Armco iron single crystal this range may, for example, be from G = 0.87 E to as low as G = 0.211 E, or for a brass single crystal from G = 1.254 E to G = 0.124 E.

Agreement with the median relationship G = 0.387 E is not necessarily proof of elastic isotropy since certain special orientations may give it; but disagreement is proof either of elastic anisotropy or of presence of seams or defects in the material. Table II

Table II

Material	E × 10^6	G × 10^6	G/E
0.67 per cent carbon heat treated steel wire	30.01	11.25	0.375
	30.17	11.15	0.371
Hard drawn spring wire	28.8	11.4	0.396
Music wire	29.92	11.35	0.379
	29.77	11.25	0.378
Brass wire, as drawn	14.3	5.05	0.353
Relief annealed	15.4	5.18	0.336
Phosphor bronze	15.58	6.3	0.404
Nickel	33.4	11.2	0.335
17SRT aluminum alloy	10.28	3.82	0.372

Table III

Lot No.		12		13		14		15
Not Heat Treated before coiling:								
Spring No.		1	2	3	4	5	6	7
Dynamic Method	G	9.44	9.17	9.74	9.23	9.16	9.25	9.17
	E	26.85	28.11	26.34	27.77	28.13	28.02	27.87
	G/E	0.351	0.326	0.370	0.332	0.326	0.329	0.329
Static Load	G	9.64	9.14	9.72	9.27	9.10	9.17	9.13
	E	27.11	28.45	26.45	28.07	27.96	28.20	27.61
	G/E	0.356	0.321	0.368	0.330	0.325	0.325	0.331
Heat Treated at 850° F. 30 min., before coiling:								
Spring No.		1H	2H		4H	5H	6H	7H
Static Load	G	9.74	9.55	.....	9.19	9.19	9.26	8.87
Dynamic Method	G	9.76	9.59	.....	9.24	9.23	9.29	8.88
	E	26.83	29.40	.....	26.85	27.82	27.63	27.51
	G/E	0.363	0.326	.....	0.344	0.332	0.336	0.322

which represents data gathered in connection with an investigation sponsored by the Research Committee on Mechanical Springs of the American Society of Mechanical Engineers (8), using specially selected wire and relatively high precision methods, in all but one or two cases, shows definite evidence of this anisotropy resulting from preferred orientation.

Similar measurements made for John Chatillon and Sons on different batches of isoelastic metal, which is a modified Elinvar, are given in Table III. Here the material was in helical spring form, with accompanying straight strips, and parallel measurements of Young's modulus and of the shearing modulus were made by both static and dynamic methods. The successive lots of this metal were of very nearly the same analysis, and were all heavily cold drawn, using essentially the same steps. The G/E ratio in all cases shows the effect of the preferred orientation resulting from the cold drawing operations. In this particular case these ratios were of importance because they affected the ability to obtain a straight line load deflection curve from the springs. The variations between the results for the different springs are themselves significant, the highest modulus in shear being accompanied by the lowest modulus in bending and vice versa. This is as would be expected on the basis of the single crystal modulus relationships.

A second phase of the elastic anisotropy in metal crystals and one which has apparently been relatively unnoticed in the past, is its effect upon stress distribution in the metal. As a generalized statement, in a body of metal composed of grains of varying moduli of elasticity those grains with larger modulus would be compelled to carry proportionately larger amounts of the load, exact quantitative

relationships depending upon the texture of the metal. In a hypothetical case where the grains were arranged in columnar fashion, extending from top to bottom of the bar, all of the grains would be compelled to deform under load to exactly the same extent and the unit stress in the individual grains would be directly proportional to their moduli of elasticity in the direction of the stress. This obviously, in view of the moduli of elasticity relationships already mentioned, would result in differences in unit stress between adjacent grains which might run as high as two or even three to one. This condition is approximated in heavily cold-worked material in which the grains have been considerably elongated in the direction of the stress. On the other hand, in equiaxed material in which the grains may be considered to approximate irregular spheres in shape, the stiffer grains will tend to bed themselves in the adjacent less stiff grains so that the strain distribution is no longer uniform from grain to grain. The stiffer grains will proportionately yield elastically less than less stiff ones so that while they will be under heavier unit stresses, the ratio of unit stresses between the grains will be proportionately much less than the ratio of moduli of elasticity. However, this introduces relatively high shearing stresses along the grain sides in order to take up the differences in deflections. Further, wherever two or more stiffer grains are arranged in tandem in the direction of the stress, exceptionally high stresses will tend to develop at the irregular faces of contact between these grains. For material with fibrous or columnar structure with load acting perpendicular to the length of the fibers, the stress distribution will be more or less the same as for equiaxed material as just described. On the other hand, in sheet or similar material in which the grains have been flattened rather than elongated, with loads acting parallel to either the length or width of the grain, the tendency will be for a stress distribution proportional to modulus of elasticity precisely as in the fibrous material. At right angles to the broad face of the grain, however, we have what is essentially equivalent to a series of thin layers composed of material of differing stiffness piled one on top of another. With stress acting perpendicular to these layers obviously each grain must carry the same load per square inch as any other grain irrespective of modulus of elasticity.

Offhand, it would appear that equiaxed material would give the most satisfactory type of stress distribution. Note, however, that the plane of easiest slip for most of our commercial metals is either

the {1 1 1} or the {1 1 0} plane, and that the directions of slip in these planes are {1 1 0} or {1 1 1} respectively. These are also the planes and the directions of greatest atomic density. Slip is possible along other planes as well, but at higher values of shearing stress. These weak planes are advantageously placed to resist tensile or compressive stresses acting along the {1 1 1} direction, which is the direction of highest Young's modulus. Conversely, forces acting along the major axes of the crystal have maximum shearing components parallel to the weak {1 1 0} or {1 1 1} planes. There are therefore compensatory effects, those grains, one of whose major axes lies parallel to the direction of stress, being able to carry the least load before slip occurs, while at the same time, due to the low value of Young's modulus parallel to the major axes, they receive the least stress. Conversely, those grains whose {1 1 1} plane is perpendicular to the direction of stress have the highest stiffness and so receive the greatest stress, but are best able to carry it. These statements are, of course, qualitative rather than quantitative in character, and work pointing toward the development of quantitative relationships would be highly desirable.

Certain general conclusions from the statements made in the last few paragraphs may be drawn.

(1) The modulus of elasticity for any one metal must be thought of not as an exact constant, but rather as a variable dependent upon the texture and upon the degree of preferred orientation of the metal. For material with random grain orientation, free from flaws and from the after effects of over stress, values of moduli of elasticity in shear and torsion corresponding to those given in the usual handbook tables may be expected. The effect of over stress not followed by later stress relief will usually be to lower both E and G. On the other hand, the effect of preferred orientation will be to raise the value of one and lower that of the other, or vice versa. Where the preferred orientation results from cold work the general tendency will be so to change the crystal orientation as to raise the value of E while lowering that of G. Hard rolled bus-bar copper, for example, may have a noticeably higher modulus than the annealed bars from which it was made.

This raises the interesting speculation as to the future possibility, by control of crystal orientation coupled with maintenance of small grain size of obtaining a metal which would occasionally be highly desirable; namely a steel with satisfactory strength but with much

higher or much lower modulus than the customary 28,000,000 to 30,000,000.

(2) The directional gain in strength resulting from proper grain flow in hot or cold working is evidently in part explained by the development of more satisfactory stress distribution relationships between the grains. The suggestion is obvious that a detailed study of the mechanism by which this occurs may aid in obtaining still better results in the future.

(3) The variations in effective elastic stiffness in adjacent grains or from spot to spot in the face of a sheet may also prove to be one factor in connection with occurrence of poor surface or other erratic behavior in pressing of sheets.

(4) From the point of view of the designer, the presence of these large variations in stress from grain to grain emphasizes the already existing reasons in favor of less emphasis on ultra-refined stress computations, particularly at points of local concentration, coupled with more emphasis on the effect of flow lines in the metal in the immediate neighborhood of the stress concentration.

It is appreciated that in large measure the description here given is qualitative rather than quantitative in nature. It is given in that form partly because of absence of quantitative data and partly as a means of urging further research upon various topics here brought up.

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THE EFFECT OF THE SHAPE OF THE TEST PIECE UPON THE ENERGY NEEDED TO DEFORM MATERIALS IN THE SINGLE-BLOW DROP TEST

By Owen W. Ellis

Abstract

A study is made of the energy needed in the single-blow drop test to deform cylinders, frusta, and cones (of equal height and volume) of copper and nickel. Graphs are given, showing the relationship between percentage reduction in height of sample and ratio of smaller to larger diameter of frusta, (1) for copper samples of volume equal to that of normal 1-inch samples and (2) for nickel samples of volume equal to that of normal ½-inch samples.

The experiments on nickel show that an equation of the form $E = bD^n$, similar to that found by the author to apply in the case of straight carbon steels, relates energy of blow in the single-blow drop test and percentage reduction in height of sample. In this equation n shows a linear relation to b as follows: $n = 1.55 - 0.52 \log b$.

It has been found that this equation, or one similar to it, applies not merely to cylinders, but to cones and conic frusta of equal volume. It applies, further, to steel and nickel at all temperatures, and to copper, lead, and possibly aluminum at room temperatures. Experiments on cadmium, tin, and brass showed that the equation did not apply to them at room temperature.

The mode of deformation of cylinders, frusta, and cones is discussed in the light of etched sections of nickel samples of these various forms. A quantitative study of such etched sections has led to the tentative conclusion that the energy absorbed in the formation of the "ovoidal segment" becomes proportionally greater as the forging temperature is lowered and as the energy of the blow is increased.

IN a recent paper by the present author,[1] there has been described the results of tests made to determine the relationship between the energy of the blow in a single-blow drop test and the percentage re-

[1] O. W. Ellis, "The Malleability of Nickel and Monel Metal," *Journal*, Institute of Metals, 1934, No. 1, Vol. 54, p. 145-160.

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duction in height of cold-drawn nickel test pieces ½ inch in diameter and ½ inch high—so-called "normal half-inch test samples."

On halving certain of the deformed nickel samples and etching the polished sections with nitric-acetic acid etching solution, the author found that the effects of deformation upon the macrostructure of the pieces were nicely disclosed. He determined, therefore, to study further the effect of shape upon the energy needed to deform samples of various materials in the single-blow drop test, and at the same time to determine, if possible, the mode of deformation of the test pieces.

The author had already (1931) investigated the effect of shape upon the energy needed to deform test pieces of copper at room temperature. These pieces had all been removed from a single rod of annealed, high-conductivity copper. They were in the form of right cylinders, frusta of cones, and cones—all of equal volume.

The dimensions and weights of the pieces are given in Table I.

Table I
Dimensions and Weights of Test Pieces of Annealed Copper

Shape of Test Piece	Height of Test Piece (in.)	Smaller Diameter of Test Piece (in.)	Larger Diameter of Test Piece (in.)	Weight of* Test Piece (g.)
Right cylinder	1.00	1.00	1.00	113
Frustum	1.00	0.80	1.19	115
Frustum	1.00	0.60	1.35	117
Frustum	1.00	0.40	1.50	117
Frustum	1.00	0.20	1.61	114
Cone	1.00	...	1.73	117

(*Weights quoted are averages of four test pieces.)

The general appearance of a series of test pieces is shown in Fig. 1. Such a series, one of each shape, was subjected to a single blow of 169 foot-pounds, another to a single blow of 282 foot-pounds, another to a single blow of 394 foot-pounds, and a fourth to a single blow of 508 foot-pounds. The heights of the samples were measured

Table II
Percentage Deformation of Test Pieces of Various Shapes under Blows of Different Energy Contents (Annealed Copper)

Shape of Sample	Ratio of Smaller to Larger Diam.	Energy of Blow (Tests Made at Room Temp.) 169 Ft. Lb.	282 Ft. Lb.	394 Ft. Lb.	508 Ft. Lb.
Right cylinder	1/1	5.5%	8.6%	11.0%	14.2%
Frustum	0.80/1.19	6.4%	9.6%	12.7%	16.2%
Frustum	0.60/1.35	8.5%	11.5%	15.8%	17.5%
Frustum	0.40/1.50	11.4%	15.9%	19.2%	22.5%
Frustum	0.20/1.61	18.4%	22.8%	27.0%	30.0%
Cone	0/1.73	26.0%	31.4%	35.0%	37.2%

after deformation, and their percentage reductions in height calculated. Results of these single-blow drop tests are shown in Table II.

These results are reproduced graphically in Fig. 2, wherein the

Fig. 1—Diagrammatic Representation of Shapes of Test Pieces Used in the Experiments. All Samples Were of the Same Volume.

percentage reductions in height of the test pieces are plotted against the ratios of the smaller to the larger diameters of the test pieces.

In the experiments on nickel, smaller test pieces were employed. The nickel rod used for the test pieces had been annealed at 1000 degrees Cent. (1830 degrees Fahr.). The volumes of the test pieces were in all cases the same (within ordinary machining limits), being equal to that of a normal ½-inch sample. In order to check the volumes of the test pieces they were weighed after machining. The

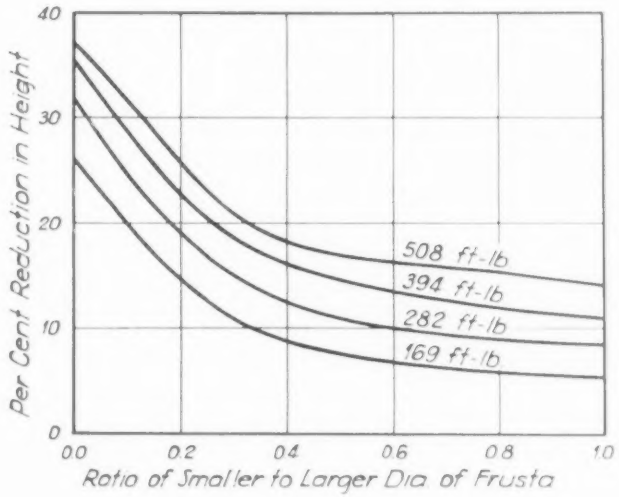


Fig. 2—Percentage Deformations of Annealed Copper Test Pieces of Various Shapes Under Blows of Different Energy Contents.

pieces were all ½ inch high. The dimensions of the pieces were so chosen that the ratios of the smaller to the larger diameters of the right cylinders, frusta, and cones were 1.0, 0.8, 0.6, 0.4, 0.2, and 0, respectively.

The pieces were forged at temperatures of 600 degrees Cent. (1112 degrees Fahr.), 800 degrees Cent. (1472 degrees Fahr.), and 1000 degrees Cent. (1830 degrees Fahr.), under blows of 75.3, 301.3, and 527.3 foot-pounds, respectively. Prior to forging, they were held in a controlled furnace at each temperature for 10 minutes, part of which time was occupied by the pieces in reaching their maximum temperature. The heights of the pieces were measured after forging, and their percentage reductions in height later calculated. The results of these tests are given in Table III.

Table III
Percentage Deformation of Test Pieces of Various Shapes under Blows of Different Energy Contents at Different Temperatures
(Annealed Nickel)

Shape of Test Piece	Ratio of Smaller to Larger Diam.	Temp. of Forging (°C.)	Energy of Blow 75.3 Ft. Lb.	301.3 Ft. Lb.	527.3 Ft. Lb.
	1.0	600	8.4*	30.0*	44.4*
Right cylinder	1.0	800	12.6*	39.2*	55.6*
	1.0	1000	21.0*	51.2*	65.2*
	0.8	600	15.5	36.9	50.7
Frustum	0.8	800	17.5	42.4	56.7
	0.8	1000	21.9	51.5	65.2
	0.6	600	16.6	39.2	53.7
Frustum	0.6	800	19.3	43.7	58.6
	0.6	1000	23.9	53.2	65.9
	0.4	600	19.4	42.4	56.8
Frustum	0.4	800	22.4	49.5	60.4
	0.4	1000	26.8	55.8	67.1
	0.2	600	26.2	50.5	60.4
Frustum	0.2	800	30.0	55.7	67.1
	0.2	1000	34.8	62.3	72.5
	0	600	37.1	57.5	65.8
Cone	0	800	40.9	62.0	71.6
	0	1000	44.5	68.4	76.0

*These values cannot be compared with those quoted elsewhere in the table. They are the results of tests made on cold-drawn (not annealed) nickel test pieces (see paper on "The Malleability of Nickel and Monel Metal," *Journal*, Institute of Metals, 1934, Vol. 54, p. 145-160).

They are presented graphically in Figs. 3-8. In Figs. 3-5 the curves show the relationship between the energy of blow in foot-pounds and the percentage deformation of the test pieces (cylinders, frusta, and cones) at the temperatures 600, 800, and 1000 degrees Cent. (1110, 1470, and 1830 degrees Fahr.), respectively. In Figs. 6-8 the curves show the relationship between the percentage deformation of the test pieces (cylinders, frusta, and cones) and the ratio of the smaller to the larger diameters of the test pieces under blows

of 75.3, 301.3, and 527.3 foot-pounds, respectively. In Figs. 6-8 the curves are drawn to fit the results obtained on the annealed samples only. The actual results obtained on the cold-drawn cylindrical pieces (see Table III) are indicated in these figures by open semicircles. The effects of initial cold work on the malleability of the cylinders forged at 600 degrees Cent. (1110 degrees Fahr.) and 800 degrees Cent. (1470 degrees Fahr.), respectively, will be understood if note be made of the positions of these semicircles relative to the

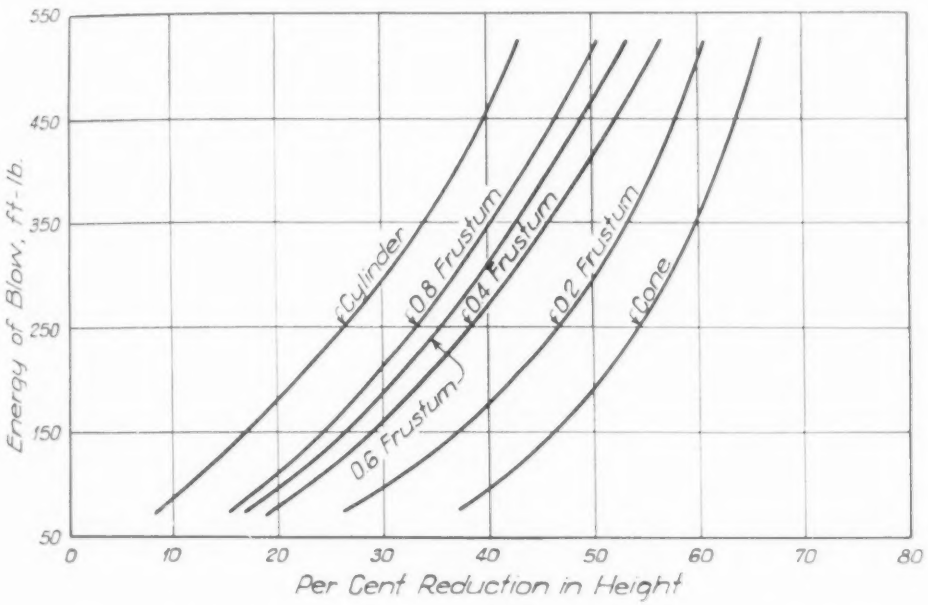


Fig. 3—Relationship Between Energies of Blow and Percentage Deformations of Annealed Nickel Test Pieces of Various Shapes, Forged at 600 Degrees Cent. (1112 Degrees Fahr.).

extrapolated curves in Figs. 6 and 7. The effects of initial cold work were completely obliterated by the mere heating of the cylindrical test pieces to the forging temperature in the case of those forged at 1000 degrees Cent. (Fig. 8).

The curves in Figs. 3-5 may be considered in conjunction with those in Fig. 9 (reproduced from the author's paper on "The Malleability of Nickel and Monel Metal"). This shows the relation between the energy of blow and the percentage reduction in height of ½-inch normal samples of cold-drawn nickel rod forged at different temperatures.

Before discussing these curves, it should be pointed out that in

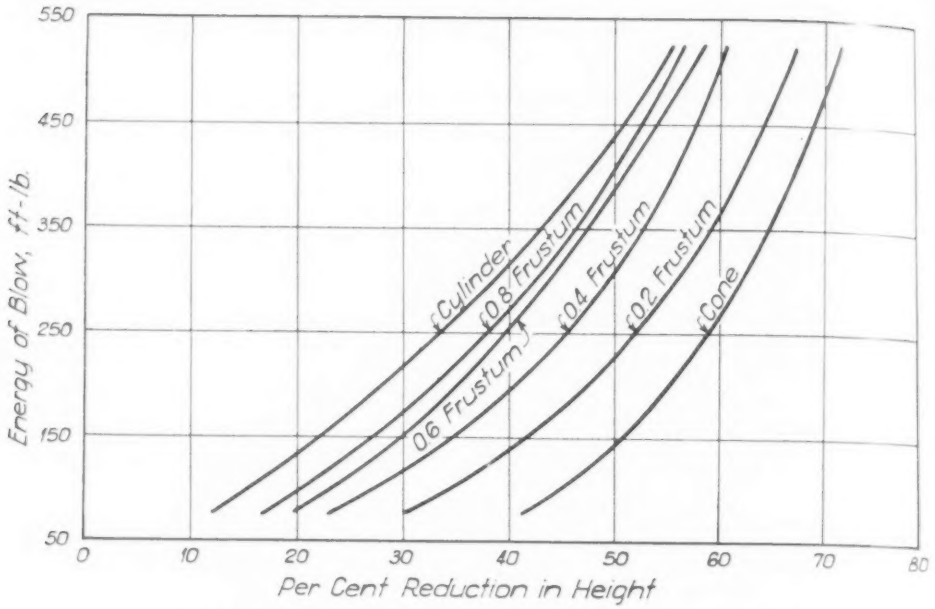


Fig. 4—Relationship Between Energies of Blow and Percentage Deformations of Annealed Nickel Test Pieces of Various Shapes, Forged at 800 Degrees Cent. (1472 Degrees Fahr.).

a paper presented to this Society,[2] the author showed that a simple equation of the form

$$E = bD^n$$

applied approximately to families of curves similar to the above in the case of straight carbon steels of various compositions. He found also that the number n showed a linear relation to b —

$$n = 1.56 - 0.47 \log b$$

so that the steel equation became

$$E = bD^{(1.56 - 0.47 \log b)}$$

In this equation, E represents the energy in foot-pounds required to produce a percentage reduction in height D of a normal sample, and b is a parameter whose value can be found by making a single drop test. In arriving at this equation, some latitude was allowed when plotting the log-log curves. In many cases the observed values for the logarithms of E and D lay close to the straight line drawn through the entire set of observations. Occasionally, however, individual

[2]O. W. Ellis, "Further Experiments on the Forgeability of Steel," TRANSACTIONS, American Society for Steel Treating, Vol. 21, 1933, p. 673-699.

points lay well away from the chosen line; this was particularly true of the lower values for E and D.

An equation of the form

$$\frac{E}{D} = bD^{n-1} + k,$$

more closely agreeing with all the experimental results, might have been deduced. Such an equation has the advantage that E/D need

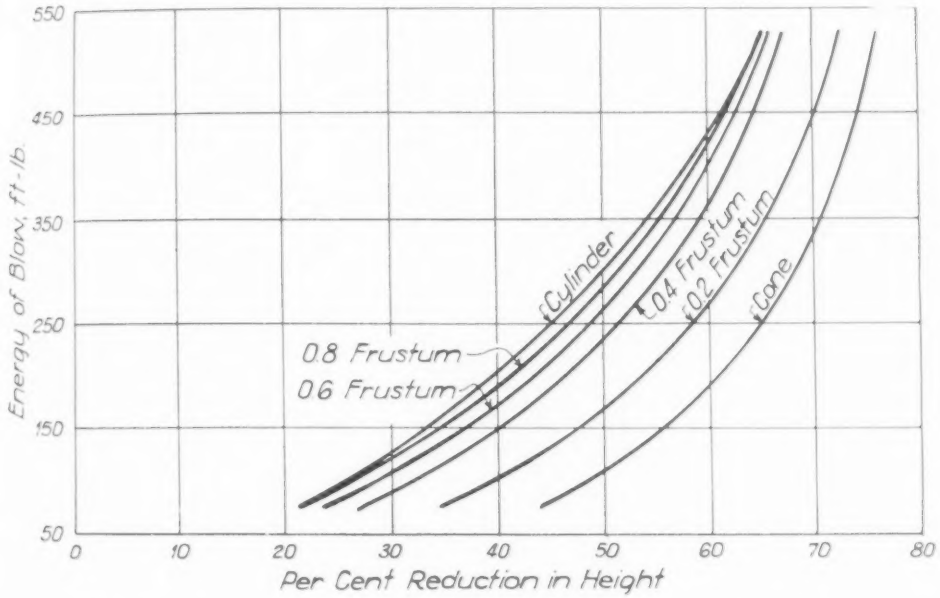


Fig. 5—Relationship Between Energies of Blow and Percentage Deformations of Annealed Nickel Test Pieces of Various Shapes, Forged at 1000 Degrees Cent. (1832 Degrees Fahr.).

not vanish with D. Its application, however, would have been limited on account of its complexity.

In discussing the author's paper on the forgeability of steel, Main pointed out that the parameter b was equivalent to the energy E needed to produce a 1 per cent reduction in height D of a normal ½-inch sample. He noted also that "a formula of the same general form applies approximately in the case of copper cylinders tested cold."

For steel in the form of ½-inch normal samples, then, the equation relating energy of blow to percentage deformation reads

$$E = bD^{(1.56 - 0.47 \log b)}.$$

Turning once again to the nickel curves (Figs. 3-5), a formula of the type

$$E = bD^n$$

has been found to apply. In this, n is related to b by the equation

$$n = 1.55 - 0.52 \log b,$$

so that the nickel equation becomes

$$E = bD^{(1.55 - 0.52 \log b)}.$$

Of special interest, however, is the fact that the nickel formula applies, not merely to cylinders, but to cones and conic frusta of equal

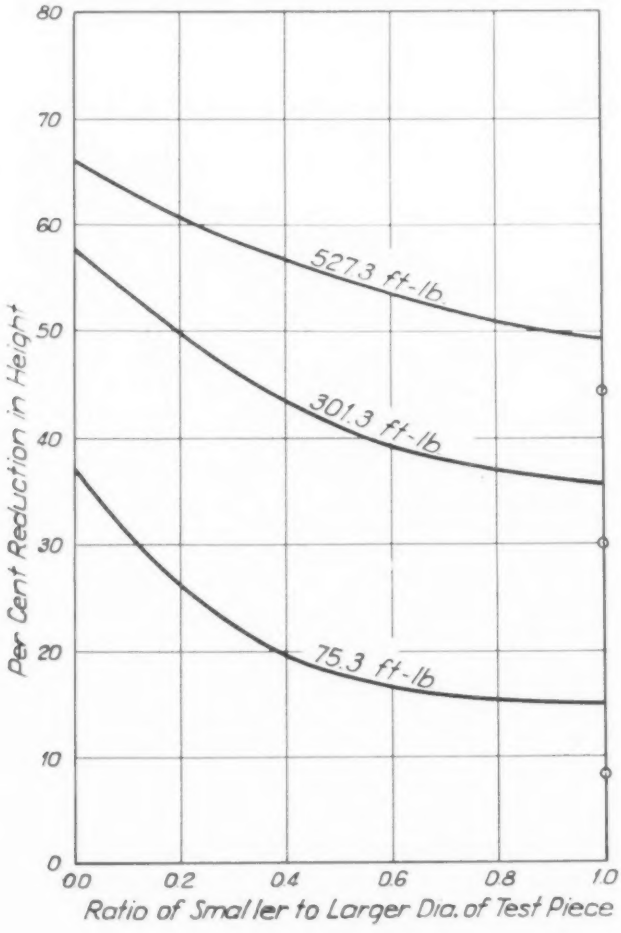


Fig. 6—Relationship Between Percentage Deformations of Annealed Nickel Test Pieces of Various Shapes and Ratios of Larger to Smaller Diameters of Test Pieces—Forged at 600 Degrees Cent. (1112 Degrees Fahr.).

volume (see Table III) as well. It should be noted that the steel equation

$$E = bD^{(1.56 - 0.47 \log b)}$$

gives results with nickel not far different from those obtained with the nickel equation.

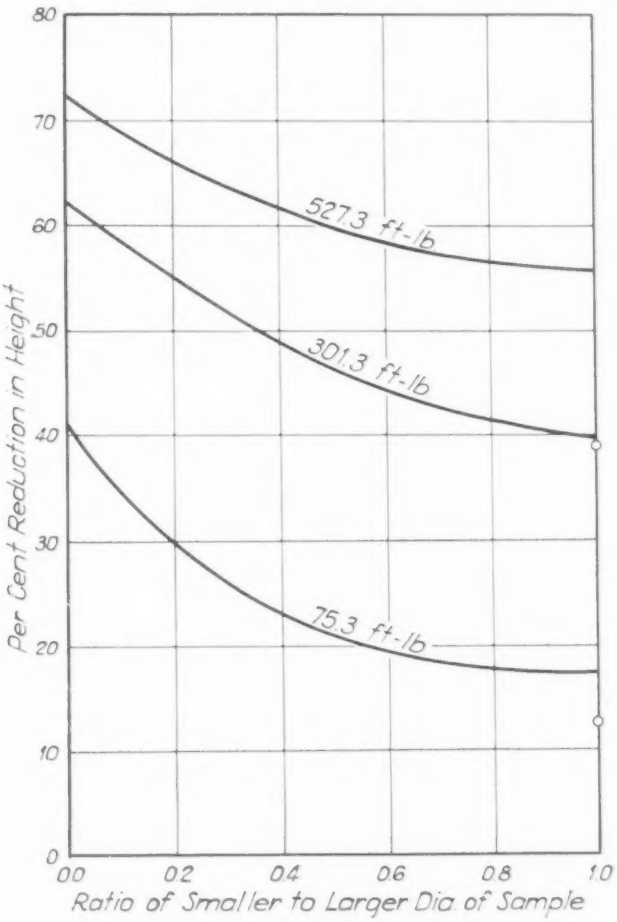


Fig. 7—Relationship Between Percentage Deformations of Annealed Nickel Test Pieces of Various Shapes and Ratios of Larger to Smaller Diameters of Test Pieces—Forged at 800 Degrees Cent. (1472 Degrees Fahr.).

Main's remarks regarding copper cylinders, quoted above, led the author to experiment with a few ½-inch normal cylinders of this material with the result that the steel equation was found to apply quite satisfactorily. For example, a normal ½-inch sample of copper was reduced 36.7 per cent in height at room temperature under a

blow of 188.3 foot-pounds. In this case the logarithm of the parameter b was found to be —0.6274. By using this parameter in the steel equation, it was estimated that to reduce a similar sample of copper 64.1 per cent in height would require an expenditure of 529.9

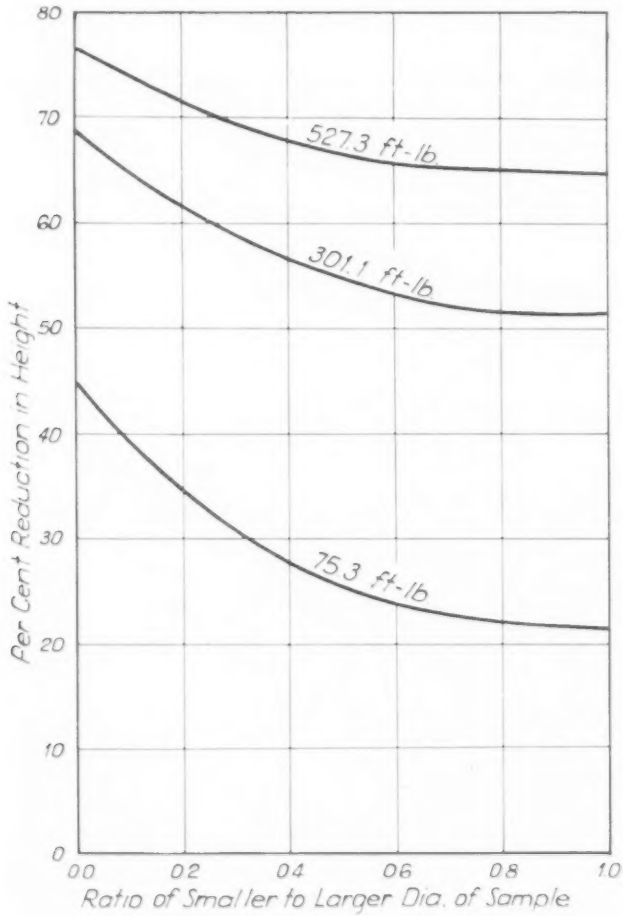


Fig 8—Relationship Between Percentage Deformations of Annealed Nickel Test Pieces of Various Shapes and Ratios of Larger to Smaller Diameters of Test Pieces—Forged at 1000 Degrees Cent. (1832 Degrees Fahr.).

foot-pounds of energy; actually a blow of 527.3 foot-pounds was needed.

The steel equation was also used to estimate the energy required to deform a ½-inch normal sample of cast lead. In this case 75.3 foot-pounds was absorbed in reducing the height of one sample by 74.4 per cent. The logarithm of the parameter b was found to be —8.6692. By using this value in the steel equation, it was esti-

mated that 195.2 foot-pounds would be absorbed in the reduction of a similar sample by 88.1 per cent in height; the energy actually needed to obtain this result was 188.3 foot-pounds.

Neither the steel nor the nickel equation gave good results in the case of aluminum. Given that a ½-inch normal sample of aluminum

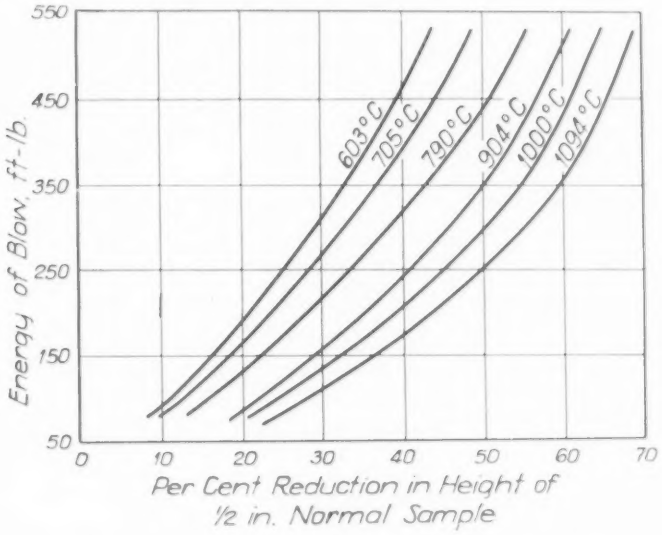


Fig 9—Relationship Between Energies of Blow and Percentage Reductions in Height of ½-inch Normal Samples of Cold-Worked Nickel Test Pieces.

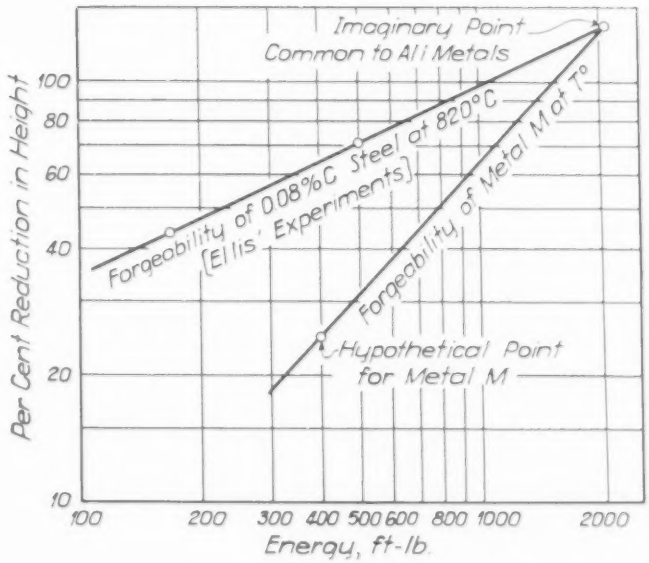


Fig 10—Graph on Log/Log Coordinates, Showing Relationship Between Energies of Blow and Percentage Reductions in Height of Normal ½-inch Samples of Various Metals (According to Main).

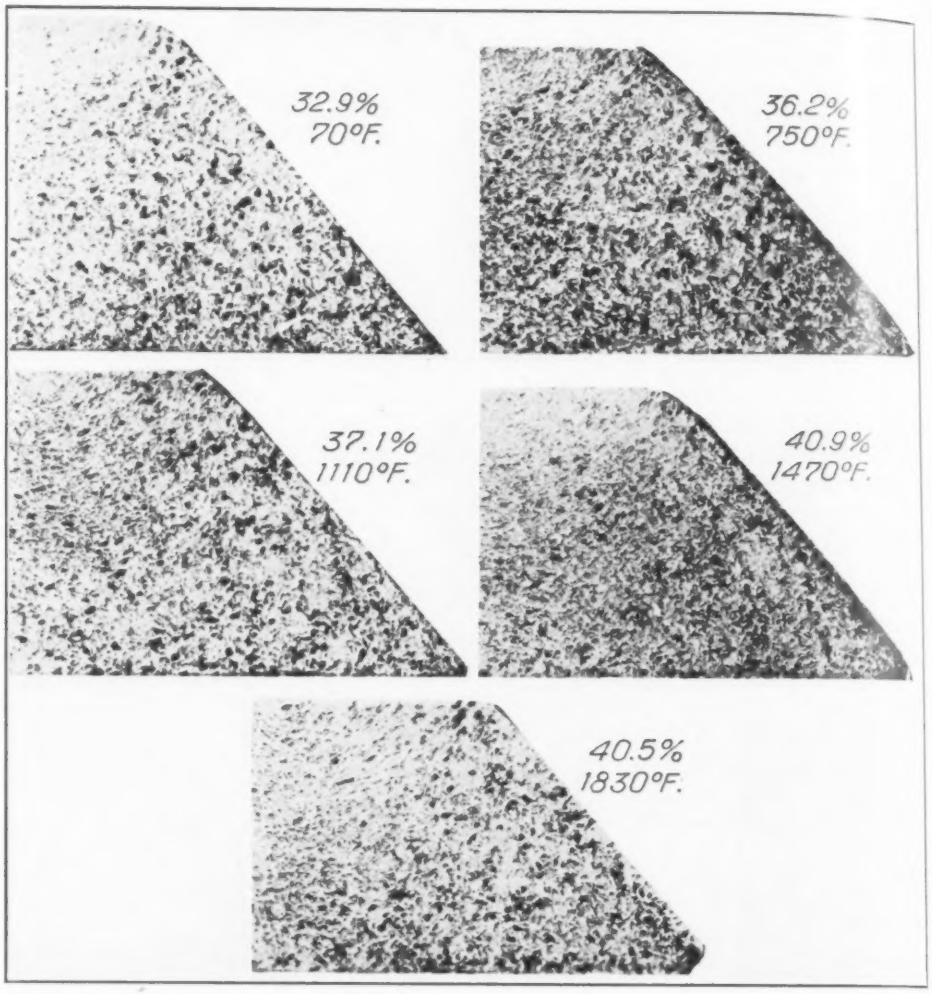


Fig 11—Etched Half Mid-Sections of Conical Samples of Annealed Nickel, Forged at Various Temperatures Under Single Blows of 75.3 Foot-Pounds.

was reduced 38.4 per cent in height under a blow of 188.3 foot-pounds, it was estimated (by using the steel equation) that a similar sample would require a blow of 572.9 foot-pounds to reduce it 68.5 per cent in height, whereas a blow of only 527.3 foot-pounds was needed. The estimate was 8.7 per cent too high.

Still less satisfactory results were obtained with the steel equation in tests on cadmium, tin, and a brass of unknown composition. The brass consisted essentially of alpha solid solution (about 90 per cent). The microconstituents in the rod were drawn out in a longitudinal direction, indicating clearly that the brass had been the subject of

plastic deformation during manufacture. The results of these tests are given in Table IV.

Table IV
Results of Single-blow Drop Tests on Brass, Cadmium, and Tin

Alloy or Metal	Experiment I		Experiment II		Estimated Energy of Blow, Using Steel Formula and Results of Experiment I
	Energy of Blow	Reduction of Height	Energy of Blow*	Reduction of Height	
	(ft. lb.)	(%)	(ft. lb.)	(%)	(ft. lb.)
Brass	188.3	26.0	527.3	49.6	445.9
Cadmium (cast)	75.3	26.4	188.3	51.2	291.4
Tin (cast)	75.3	39.5	188.3	68.3	349.4

*Actual energy to be compared with estimated energy (column 6).

Insofar as steel and nickel (at all temperatures) and copper lead, and possibly aluminum (at room temperature) are concerned, the above results support the view that an equation of the form

$$E = bD^{(p - q \log b)}$$

represents with reasonable accuracy the relationship between energy of blow E and percentage deformation D of samples (of various simple geometrical shapes—cylinders, frusta, and cones) in the single-blow drop test.* It seems not improbable, though most surprising, that the values p and q in the above equation are in the close neighborhood of 1.56 and 0.47, respectively, for *all* these materials under the conditions specified.

As Main has shown, the equation

$$E = bD^{(1.56 - 0.47 \log b)}$$

can be rewritten

$$3.32 - \log E = a(2.13 - \log D)$$

where a, the parameter chosen in this equation, is equal to (1.56 — 0.47 log b). In view of their importance, Main's further remarks in this connection are quoted below.

> "In the imaginary condition when log D = 2.13 (corresponding to a reduction of 134.9 per cent), log E is 3.32 (E = 2089 foot-pounds) whatever the value of a, for then the right side of the second equation vanishes and log E becomes equal to 3.32. That means that under these conditions

*This equation is unsatisfactory for low values of E and D.

all materials coming within the formula at all temperatures would behave alike.

"These conditions can only exist in a mathematical sense, but this fact assists the plotting, for on double logarithmic paper, E and D give a straight line, the slope of which is the parameter a.

"All such lines for all materials and for all temperatures covered by the formula pass, when produced beyond the D = 100 per line, through the point D = 134.9, E = 2089 foot-pounds. This is shown on the accompanying diagram.* It follows that all that is necessary to determine the forgeability of a normal sample of steel at a given temperature is to join up any single accurate observation with this point by a straight line, which shows all corresponding values of deformation and energy for that particular steel and temperature.

"The divergence downwards of such lines indicates that when the behavior of any steel at any temperature is being compared with that of the same or any other steel, at the same or any other temperature, the ratio of difference becomes less the larger the amount of deformation on which the comparison is based."

What Main says of normal samples of steel seems likely, from the new experiments on nickel, to apply with equal force to samples of various shapes (cylinders, frusta, and cones) of the metals, aluminum (?), copper, lead, and nickel. There is some justification, therefore, for Main's use of the phrase "Metal M" in his diagram reproduced as Fig. 10 hereof.

It is certainly matter for surprise that one such simple equation as the above should represent with reasonable accuracy for no less than five materials (and one of these carbon steel in the broad sense) the relationship between energy of blow and percentage deformation of a sample in the single-blow drop test; and, further, that a single parameter should serve so completely to define this relationship at any given temperature. This parameter, in fact, represents in the equation all that is characteristic of the metal or of its malleability. It might, therefore, be appropriately named the "malleability number." In view of the foregoing, however, the fact that this equation fails when applied to brass, cast cadmium and cast tin is matter for surprise, also. The question naturally arises whether crystal structure,

*Reproduced on page 953 as Fig. 10.

grain size, or crystal orientation accounts for the marked difference between the actual energy and the estimated energy (by using the steel equation to find the parameter b, and so on) required to deform these materials. The longitudinal orientation of the two phases, alpha and beta, in the brass rod used in these tests might possibly explain why the actual energy (527.3 foot-pounds) required to reduce a sample by 49.6 per cent in height exceeded the estimated energy (445.9 foot-pounds) by more than 18 per cent. Had the castings from which the cadmium and tin samples were removed been characterized by columnar crystals radiating from the centers to the circumferences of the castings, the peculiar behavior of the samples might possibly have been accounted for by the favorable orientation of densely populated slip planes within the columnar crystals relative to the direction of the applied stresses during forging; but, while the crystals in both the cadmium and the tin castings were large relative to those in the samples of aluminum, copper, lead, nickel, and steel, they were, nevertheless, equiaxed.

It is hard, too, to believe that grain size of these metals can explain their peculiar behavior, since the grain sizes of both the nickel and the steel samples varied quite widely on account of the ranges of forging temperatures employed in the tests on these materials. Further, the grain size of the cast lead, the relation between whose deformation and energy absorption was satisfactorily represented by the steel equation, was of the same order as that of the cadmium and the tin.

One is left, then, with the crystal structure to explain the inconsistencies of these metals. However, the author finds it difficult at the moment to believe that crystal structure alone can account for effects which are due to movements within, and of, large numbers of crystals—effects which are essentially statistical in nature.

The author has mentioned that etching the polished sections of a number of deformed nickel cylinders with a suitable reagent brought out quite nicely the effects of deformation on their macrostructure. This observation led him to enquire into the mode of deformation of some of the simpler shapes which were the subject of his tests (Table III).

It is well known that in the compression testing of ductile, cylindrical samples the material undergoing maximum distortion is more or less concentrated within two cup-shaped surfaces, these surfaces enclosing what may be described as ovoidal segments. The

bases of these ovoidal segments correspond to the upper and lower surfaces of the compressed cylinders; the apexes intermingle, rather than meet, in the neighborhood of the centers of the compressed cylinders.

In the compression testing of cones the material undergoing distortion is more or less concentrated within *one* cup-shaped surface. The base of the ovoidal segment in this case corresponds to the smaller circular surface of the frustum which results from the compression of the cone; the apex of the segment lies in the direction of the larger circular surface of the frustum, and may rest within or without the frustum according to the amount of deformation to which the cone has been subjected. The conditions existing in conical frusta during and after deformation in the single-blow drop test are intermediate in character between those in cylinders and cones.

These and other facts can be deduced from an examination of suitably etched sections of forged nickel shapes. A series of such sections, halved for convenience of reproduction, is shown in Fig. 11. Of these, the uppermost had been reduced 32.9 per cent in height under a blow of 75.3 foot-pounds at room temperature; the next, 36.2 per cent in height under a similar blow at about 400 degrees Cent. (750 degrees Fahr.); the next 37.1 per cent in height under a similar blow at 600 degrees Cent. (1110 degrees Fahr.); the next 40.9 per cent in height under a similar blow at 800 degrees Cent. (1470 degrees Fahr.); and the last, 44.5 per cent in height under a similar blow at 1000 degrees Cent. (1830 degrees Fahr.).

A study of the uppermost section under the microscope showed it to be divisible into two sharply defined zones, as in Fig. 12a. The smaller zone (a section of what has been termed above an "ovoidal segment") was characterized by well-deformed grains, the larger by equiaxed crystallities. The deformed grains in the smaller zone were stratified, the strata lying one within the other like the scales of an onion.

The grains in the larger zone showed no signs of deformation, though the metal must have suffered some radial displacement and circumferential strain. This circumferential strain was in all probability of the order of half the strain in the ovoidal segment.

The further progress of the deformation of a cone is illustrated in Figs. 12b and 12c. The frustum, a mid-section of which is shown in Fig 12b, was formed as a result of a 55 per cent reduction in height at room temperature of a cone, under a blow of 301.3 foot-

pounds. That shown in Fig. 12c was produced by a 65.3 per cent reduction in height at room temperature of a cone, under a blow of 527.3 foot-pounds. In each case, the mid-section of the frustum is divisible into three sharply defined zones, one being a section of an ovoidal segment whose base corresponds with the top of the frustum, and whose apex lies below the base of the frustum. The other zones are sections of an annulus.

In each case, the deformed grains within the ovodial segment

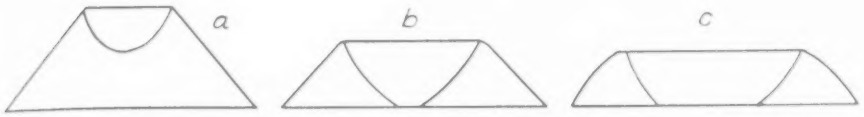


Fig. 12—Diagrammatic Representation of Appearance of Etched Mid-Sections of Cones of Nickel, Forged Under Blows of Different Energy Contents.

were stratified, the strata lying one within the other like the scales of an onion. Having been cut at right angles to their longer axes, the grains in the annulus showed little or no sign of deformation. The boundary between the ovoidal segment and the annulus was thus clearly defined. The metal forming the annulus had, of course, suffered considerable radial displacement and circumferential strain as a result of the piercing and expanding actions of the ovodial segment, which, during deformation, must have been growing in size.

Had greater care been taken in the bisection of these and the many other samples which were examined by the author during this investigation, some ideas might have been gained regarding the manner in which the energy of the blow was distributed during forging—what proportion, for example, was absorbed in the formation of the ovoidal segment in each case, and what proportion was taken up in piercing and expanding the surrounding annulus. Unfortunately, the thought that some such relationships might be developed did not occur to the author until after the samples had been cut. However, the results of such measurements as have been made on the various samples tested are quoted in Table V for what they are worth, it being realized that the estimated value of the ovoidal segments may be incorrect (generally on the low side), due, in the main, to the fact that the sections were not exactly diametrical.

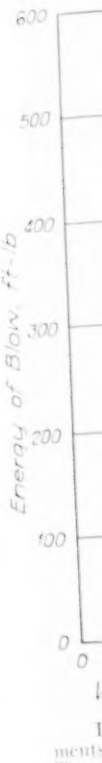
A tentative conclusion, based upon a study of the results quoted in Table V, is that, other things being equal, the energy absorbed in the formation of the ovoidal segment becomes proportionally

Table V

Estimated Volumes of Ovoidal Segments in Samples of Various Shapes

Form of Sample	Temperature of Forging	Energy of Blow (E)	Reduction in Height	Estimated Volume of Ovoidal Segment	Volume of Ovoidal Segment (V)	E/V
	(°C.)	(ft. lb.)	(%)	(cc.)	(% of Total Vol.)	
	R.T.	75.3	32.9	0.095	5.9	12.8
	100	75.3	32.9	0.108	6.8	11.1
	400	75.3	36.2	0.123	7.6	9.9
Cone	600	75.3	37.1	0.176	11.0	6.9
	800	75.3	40.9	0.229	14.3	5.3
	1000	75.3	44.5	0.268	16.7	4.5
	R.T.	301.3	55.0	0.305	19.0	16.3
	100	301.3	55.5	0.418	26.0	11.6
	400	301.3	58.7	0.393	24.4	12.4
Cone	600	301.3	57.5	0.620	38.5	7.8
	800	301.3	62.0	0.543	33.8	8.9
	1000	301.3	68.4	0.792	49.2	6.1
	R.T.	527.3	65.3	0.617	38.3	13.8
	100	527.3	65.2	0.643	39.9	13.2
	400	527.3	67.2	0.616	38.3	13.8
Cone	600	527.3	65.8	0.835	51.9	10.2
	800	527.3	71.6	1.144	71.1	7.4
	1000	527.3	76.0	1.081	67.2	7.9
	600	75.3	26.2	0.142	8.9	8.5
0.2 Frustum	800	75.3	30.0	0.197	12.2	6.2
	1000	75.3	34.8	0.202	12.5	6.0
	600	301.3	50.5	0.409	25.4	11.9
0.2 Frustum	800	301.3	55.7	0.745	46.3	6.5
	1000	301.3	62.3	0.854	53.1	5.7
	600	527.3	60.4	0.736	45.7	11.5
0.2 Frustum	800	527.3	67.1	1.309	81.4	6.5
	1000	527.3	72.5	1.120	69.6	7.6
	600	75.3	19.4	N.O.	N.O.	N.O.
0.4 Frustum	800	75.3	22.4	0.173	10.8	6.9
	1000	75.3	26.8	0.231	14.3	5.3
	600	301.3	42.4	0.405	25.2	12.0
0.4 Frustum	800	301.3	49.5	1.191	74.0	4.1
	1000	301.3	55.8	1.000	62.2	4.8
	600	527.3	56.8	0.897	55.7	9.5
0.4 Frustum	800	527.3	60.4	1.468	91.3	5.8
	1000	527.3	67.1	1.217	75.7	7.0
	600	75.3	16.6	N.O.	N.O.	N.O.
0.6 Frustum	800	75.3	19.3	N.O.	N.O.	N.O.
	1000	75.3	23.9	0.235	14.6	5.2
	600	301.3	39.2	0.352	21.9	13.8
0.6 Frustum	800	301.3	43.7	0.682	42.4	7.1
	1000	301.3	53.2	1.457	90.5	3.3
	600	527.3	53.7	0.951	59.1	8.9
0.6 Frustum	800	527.3	58.6	1.591	98.9	5.3
	1000	527.3	65.9	1.560	97.0	5.4
	600	75.3	15.5	N.O.	N.O.	N.O.
0.8 Frustum	800	75.3	17.5	N.O.	N.O.	N.O.
	1000	75.3	21.9	0.410	25.5	3.0
	600	301.3	36.9	0.432	26.8	11.2
0.8 Frustum	800	301.3	42.4	1.435	89.8	3.4
	1000	301.3	51.5	1.528	95.0	3.2
	600	527.3	50.7	0.721	44.8	11.8
0.8 Frustum	800	527.3	56.7	1.478	91.9	5.7
	1000	527.3	65.2	1.598	99.3	5.3

(N.O.—No observation. R.T.—Room Temperature.)



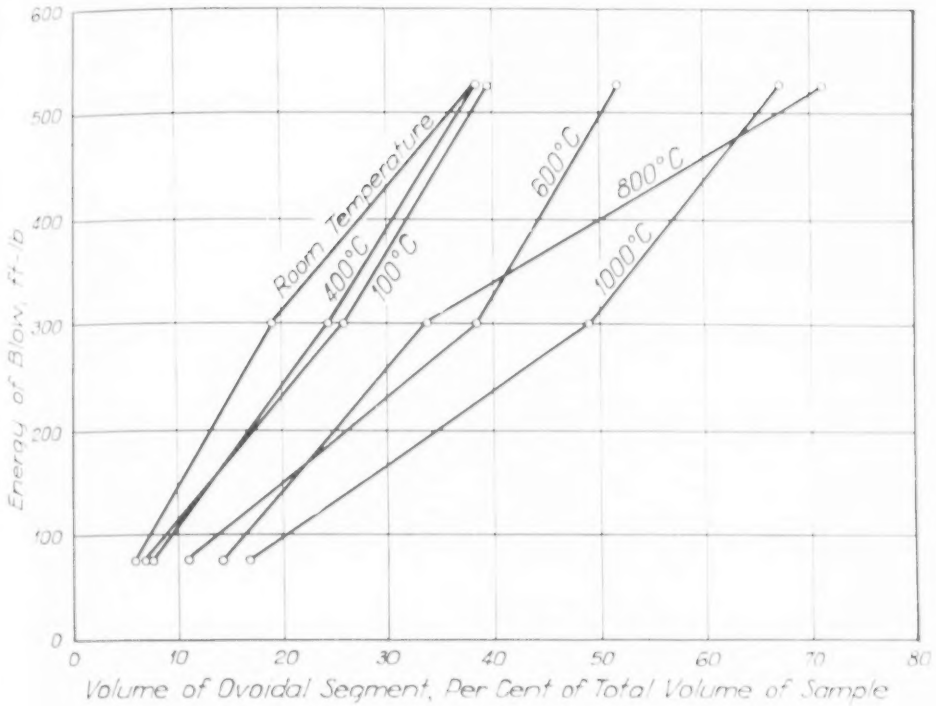


Fig. 13—Relationship Between Energies of Blow and Volumes of Ovoidal Segments (Per Cent of Total Volume of Sample) of Nickel Cones Forged at Different Temperatures.

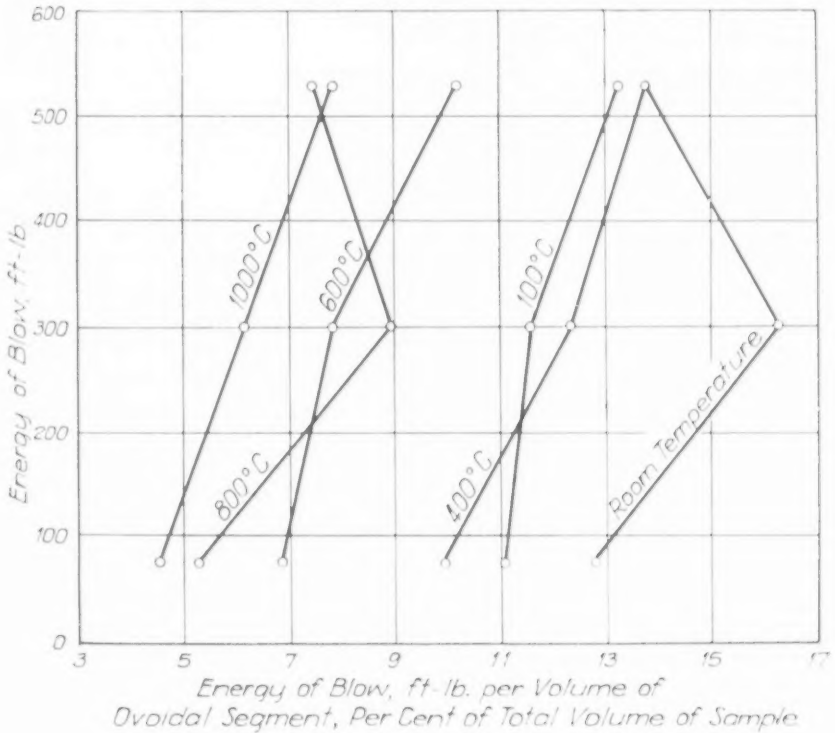


Fig. 14—Relationship Between Energies of Blow and Energies Per Unit Volume of Ovoidal Segments of Nickel Cones Forged at Different Temperatures.

greater as the forging temperature is lowered. The results quoted in Table V, insofar as they apply to cones, are shown graphically in Fig. 13, wherein the energy of the blow in foot-pounds is plotted against the volume of the ovoidal segment in a number of conical samples.

It is realized that an appreciable proportion of the energy of the blow in the single-blow drop test is absorbed in other ways than in

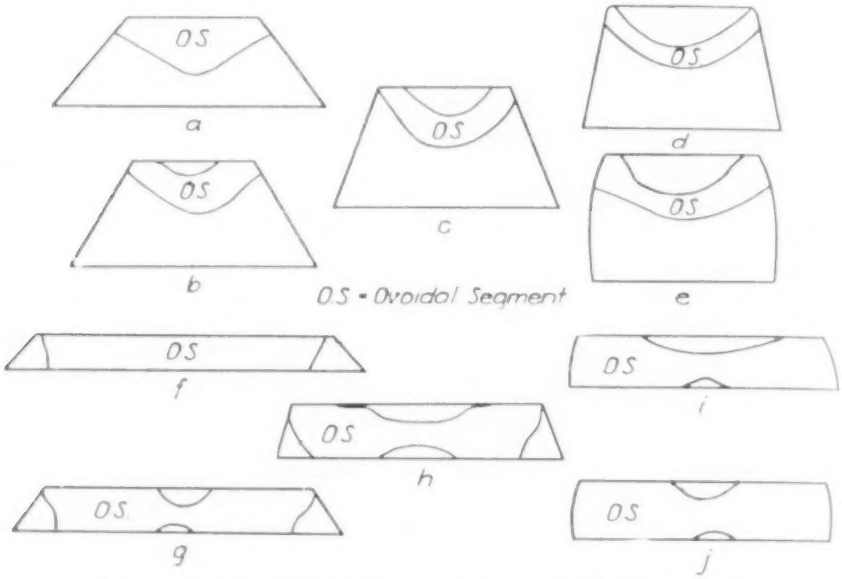


Fig. 15—Diagrammatic Representation of Mid-Sections of Samples of Various Shapes, Forged Under Blows of Different Energy Contents at 1000 Degrees Cent. (1832 Degrees Fahr.).

the deformation of the test piece (friction in the guides of the testing machine, heating of the sample due to deformation, sound resulting from the blow, etc.). Speaking generally, the energy so absorbed becomes proportionally greater as the forging temperature is lowered. It is doubtful whether the energy so absorbed varies materially from sample to sample at any given forging temperature. On the assumption that the energy so absorbed remains reasonably constant in all cases, the energy absorbed in the formation of the ovoidal segment (total energy of blow, E/ ovoidal segment, V) has been calculated in all the cases dealt with in Table V (see column 7 in this table). By plotting E/V against E, as in Fig. 14, the further tentative conclusion is reached that the energy absorbed in the formation of the ovoidal segment becomes proportionally greater as the energy of the blow is increased.

Other examples of the effects of deformation upon the macrostructure of midsections of cones and frusta are shown in Figs. 15a to 15j. Sections (a) to (e) in Fig. 15 are of a deformed cone, a deformed 0.2 frustum,* a deformed 0.4 frustum, a deformed 0.6 frustum, and a deformed 0.8 frustum, respectively. The samples of which these are the sections were forged at a temperature of 1000 degrees Cent. (1830 degrees Fahr.) under a blow of 75.3 foot-pounds (see Table V). Sections (f) to (j) are of a similar series of deformed shapes which had been forged at a temperature of 1000 degrees Cent. (1830 degrees Fahr.) under a blow of 527.3 foot-pounds (see Table V).

Sections (a) to (e) call for but little comment. Under the same conditions of forging the ovoidal segment of deformed material, which characterizes the upper portions of forged conical samples, is replaced in forged frusta by an ovoidal segment of deformed material containing within itself a segment of apparently undeformed metal. The crystals of deformed material are arranged in layers much as are those in the ovoidal segments of forged conical samples.

Sections (f) to (j) merit special note. While the annuli in these cases could be definitely distinguished from the ovoidal segments, yet within the ovoidal segments large numbers of tiny, equiaxed crystals were visible. It would appear that the conditions of deformation and temperature were such as to cause the highly worked nickel to recrystallize during or subsequent to forging, with the result that new grains were formed within those portions of the shapes which had been subjected to the greatest degree of deformation. No such effects were noted in the samples that had been forged at 800 degrees Cent. (1470 degrees Fahr.) and 600 degrees Cent. (1110 degrees Fahr.), respectively. It is probable, though not certain, that recrystallization in this case is a function of temperature alone. We say "not certain" because, for example, a cone forged at 600 degrees Cent. (1110 degrees Fahr.) under a blow of 527.3 foot-pounds was reduced 65.8 per cent in height, but showed no signs of recrystallization within the ovoidal segment. The 0.6 frustum shown in Fig. 14 (i), however, after having been reduced 65.9 per cent in height under a blow of 527.3 foot-pounds at 1000 degrees Cent. (1830 degrees Fahr.), was characterized by numerous, small equiaxed crystals within the ovoidal segment.

The question may well be asked whether some of the structures

*By a 0.2 frustum is meant one whose upper diameter is 0.2 times its lower diameter.

examined may not have been incorrectly interpreted on account of recrystallization which may have occurred during or subsequent to deformation. The author is of opinion that this question can be answered in the negative, and that his assumption that the effects of deformation upon the structures of these samples offers a satisfactory clue as to the movements which occurred within them during forging can be accepted without serious objection.

Acknowledgments

The author desires to acknowledge the interest shown in this work by Dr. H. B. Speakman, the director of the Ontario Research Foundation; the whole-hearted assistance given him by Mr. J. Barbeau, now of Delco-Remy, Ltd., St. Catharines, Ont.; the valuable criticisms offered by Dr. J. N. Goodier, of the Ontario Research Foundation; and the help afforded him by Miss L. L. Blackburn in the preparation of the manuscript.

DISCUSSION

Written Discussion: By H. H. Ashdown, 323 Fourth Avenue, Pittsburgh.

I am glad of this opportunity to congratulate Mr. Ellis on his series of practical research papers and the data contained in them should be of great interest and value to manufacturers engaged in fabricating metals. By a careful study of these papers they should indicate to them the means of greater productive and increased die life.

There is no doubt, Mr. Ellis has some definite reason for conducting these experiments in the manner he has done but on reading this paper it has occurred to me its practical value would have been enhanced had the specimens been contained within a die and made to conform to some definite predetermined shape, as in this manner there may have been less loss of energy.

At the foot of page 962, Mr. Ellis states tentatively, that the energy absorbed in the formation of the ovoidal segment becomes proportionally greater as the energy of the blow is increased. I would venture to suggest that much of this absorbed increased energy results from frictional flow of the base of the specimen on the face of the hammer block and not in general deformation of the specimen. If the specimen were retained in a die of the original diameter of the base of the specimen I think the absorption of deformation could be better computed.

On page 958, attention is directed to the structural formation in the ovoidal segment as representing scales of an onion. I would ask Mr. Ellis if this structure was common at all temperatures or only relative to the specimens heated to the higher temperatures.

If the latter, it would appear that during heating, the metal at the upper portion of the cone rapidly developed a coarse crystalline structure and that the strata lines correspond with the crystalline boundaries.

Appreciating the exactness of all Mr. Ellis' work one feels somewhat diffident in questioning his findings.

Oral Discussion

V. N. KRIVOBOK:[1] Since there are so many very interesting papers at this annual meeting of the Society, one does not feel guilty if simple lack of time prevented careful study, so essential in preparing a written discussion upon most important papers. A valuable paper by Mr. Ellis is one of those that I have in mind.

Not so long ago while performing some carefully considered and arranged tests on strength of steels at high temperatures, it was suggested to me that the approximate estimation of relative strength of the alloys, quite sufficient to separate alloys into more or less promising groups, could be obtained by a very simple method and in a much shorter time. The method suggested was essentially the same drop test, which has just been presented to our attention by Mr. Ellis. I would be much interested to know if Mr. Ellis attempted correlation between resistance to impact at high temperature and the tensile strength of the same steels at the same high temperature. It is natural to expect that the results of such correlation would be influenced by the heat treatment the steels have received and perhaps other factors somewhat less obvious. Thus it should not be expected that the correlation could be expressed by a definite formulae, but even a generalized rule would be interesting and useful.

These remarks may be taken as a friendly suggestion for further work in this very interesting field, or an expression of hope that if suggested experiments have already been performed we may have the benefit of findings.

Author's Reply to Discussion

In regard to Mr. Ashdown's remarks, our experimental work has been, as you will appreciate, quite empirical in character. So far, we have made no attempt to take into account the effect of friction between the specimen and the face of either hammer or anvil. We recognize that friction exists. We felt that in these experiments the first and most important thing was to determine what happened to samples unsupported by dies. We may be able to answer some of Mr. Ashdown's questions at a later day, I believe, but it takes quite a long time to go through a series of experiments such as we have described this afternoon.

In regard to another question raised by Mr. Ashdown, I think I can best answer this by referring to a point of which I made no mention in my presentation, but which appears in the paper as preprinted. We found that in samples that were heated to relatively high temperatures and forged, there were definite evidences of recrystallization in the ovoidal segments. However, no one could examine any such samples and mistake which was which part, or which was which zone, in the test piece.

[1]Metallurgical Engineer, Allegheny Steel Co., Brackenridge, Pa.

Referring to Dr. Krivobok's remarks, we have no information regarding the relationship between percentage deformation and tensile strength or other properties at high temperatures. In a paper which was presented to the Iron and Steel Institute in 1926, Messrs. I. G. Slater and T. H. Turner took the results of certain hardness tests made by themselves at various temperatures above the critical, and plotted them against some of the results that I had obtained in my earlier experiments on straight carbon steels. While there certainly was not a very close linear relationship between the two sets of results, there did seem to be some sort of relationship approaching the linear. It must be remembered, of course, that the tests by Messrs. Slater and Turner were carried out on materials that differed somewhat in origin and composition from those I tested. It was rather surprising, therefore, to find that there existed so close a relationship as did exist. I might hazard a guess that there is a definite relationship between the results obtained in this test and those obtained in drop hardness tests. In all probability, on that account, there is a definite relationship between results obtained in the single-blow drop test and results obtained in the short-time tensile test.

HOT WORKING, COLD WORKING AND RECRYSTALLIZATION STRUCTURE

By N. P. Goss

Abstract

This paper discusses the X-ray method used in the investigation of the changes in the internal structure of the grains, when metals are hot-worked, cold-worked and heat treated.

The paper is divided into a number of sections, each of which describes the results of the various investigations. I—Introduction. II—Evidence of imperfection in crystal structure. III—The meaning of "Radial Asterism" found on X-ray diagrams of cold-worked metals. IV—Plastic deformation of single crystals and polycrystalline aggregates. V—Preferred orientation in hot-rolled strips. Va—Preferred orientation in hot-rolled rods. VI—Effect of small and large deformations on the physical properties. VII—New orientations developed during heat treatment and their stability on further cold-rolling. VIII—Time required to remove directional properties in high carbon wires, and time required to complete recrystallization of cold-worked strip steels. IX—Directional properties imparted to metals given small and large plastic deformations. X—The effect of bending and twisting on the annealed wires.

Introduction

THE X-ray method of crystal analysis has given us a better conception of the structure of matter entirely from an atomistic point of view as it reveals internal structure which older methods failed to do. Solid crystalline matter is built of atoms, arranged on space lattices of which there are 14 types, and most of the commercially important metals have very simple lattice structures, cubic (body-centered and face-centered), and hexagonal.

In the investigation of plastic flow and recrystallization of metals and alloys, the aim is to arrive at a better understanding of the forces which bind the atoms in the space lattice of the crystal, and the

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changes which occur in the crystals when elastically and plastically deformed. The X-ray method does not explain all of the phenomena of the solid state, but it has changed our viewpoint concerning the structure of matter, metals in particular. This is a distinct and valuable contribution to our knowledge of the solid state of matter. In the field of metals, the X-ray method of crystal analysis has revealed in a striking manner the changes which take place when a metal is hot-worked, cold-worked and heat treated.

Previous to the inception of X-ray methods of crystal analysis it was difficult to explain elasticity, but now it is known that metals are composed of atoms held rigidly in a space lattice, which is essentially an open structure. In such a structure the atoms can be slightly displaced from their normal positions when stressed elastically and return to their normal positions when the elastic stresses are removed.

The orientation and the internal structure of the grains in metals and alloys influences the physical properties. The X-ray method has been of great value in helping to explain the changes in the physical properties when a metal is plastically deformed and heat treated. Elaborate experimental researches have been conducted on single crystals and polycrystalline aggregates of many of the more important metals of technical importance and the changes in the internal structure have been correlated to the physical properties.

For example, Taylor and Elam (1)[1] studied the change in the shape of a single crystal of aluminum when gradually elongated in tension. These experiments proved that gliding occurs only on the {1 1 1} family of planes, and that the phenomenon of plastic flow even in single crystals is very complex, about which a great deal remains to be learned.

The object of this paper is to show how the method of X-ray diffraction can be used in following the changes in crystal structure when metals and alloys are hot-worked, cold-worked and heat treated.

Evidence of Imperfection in Crystal Structure

In the ideal case, a crystal may be defined as an aggregation of atoms arranged on a space lattice of definite geometrical configuration repeated with great regularity in three-dimensional space. The space lattices of the metals are, fortunately simple but many metallic

[1]The figures in parentheses refer to the bibliography appended to this paper.

compounds are known to be complex. The types of space lattices found in the commercial metals are shown in Fig. 1.

The X-ray diffraction pattern of a single crystal is always characterized by a pattern having sharp spots arranged on a definite geometrical pattern having various kinds of symmetry (Fig. 2). Under the microscope the surface is structureless (when polished and given a light etch), however, deep-etching reveals a structure similar to the one shown in Fig. 3. Deep etch figures have not received the attention they are entitled to, but their value will certainly be recognized in the future. Pulsifer (2) recognized the value of deep etch figures in studying the internal structure of grains. He termed it "fine structure" and described it as follows: "The so-called grains or crystals in metals are actually sheaves or packets of much smaller units all similarly oriented, a light or passive etch hardly reveals a trace of them but they do appear on deep etching. The lack of more knowledge of the fine structure can be attributed directly to the prevailing custom of smooth mechanical polishing and faint etching."

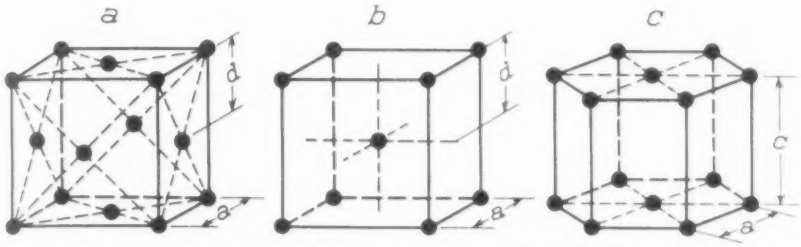
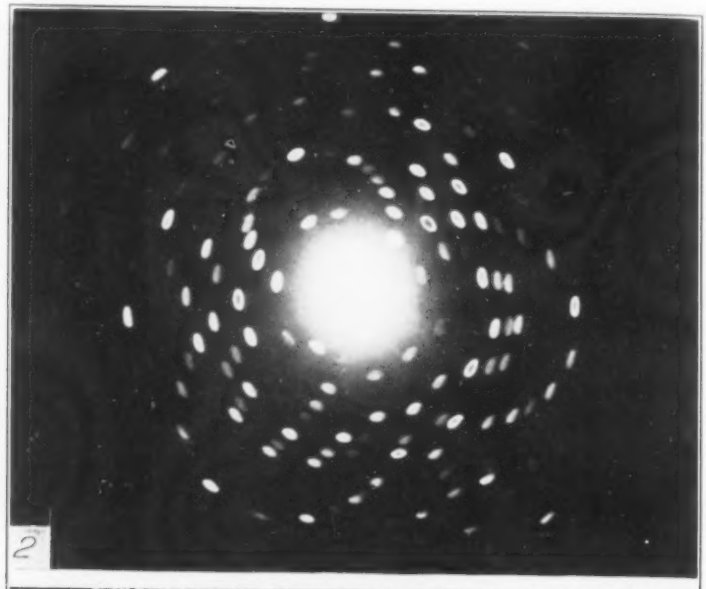


Fig. 1—Space Lattice Types. a—Face-Centered Cubic, b—Body-Centered Cubic, c—Close Packed Hexagonal.

Desch (3) devotes an entire chapter to this important phase of metallography and explains some of the results derived from such studies, along with the limitations. For example, the same face of a crystal when etched with different reagents will alter the etch figure since the crystal has directional properties. Dana Smith and Mehl (4) have shown that etching pits can be used to determine the orientation of iron crystals, in polycrystalline materials or single crystals. Another outstanding characteristic of all etch figures is their uniform size and equal distribution. This is related to periodic variations in the crystal structure. The deep etch figure (Fig. 3) of the silicon steel certainly does suggest regularity, but it is not perfect, for example, the cube edges are not quite parallel. Crystals having etch figures of this kind were examined with the X-ray, the results of this investigation will be described later in this paper.

Deep etch figures and X-ray diffraction methods will eventually advance our knowledge of the internal structure of crystals and their relationship to the physical properties. A grain of a pure metal when examined under the microscope has always been considered to be a crystal. For example, pure iron under the microscope appears as a



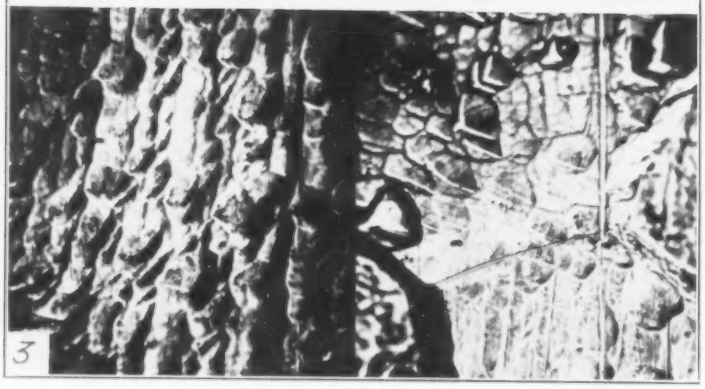


Fig. 2—X-Ray Laue Diagram of a Single Crystal of Fe_3O_4 X-Ray Beam Transmitted Parallel to (1 1 1) Direction.
Fig. 3—Deep Etched Figure of a 2 Per Cent Silicon-Ferrite Alloy Strip. Original magnification ×1500 reduced to ×520 in reproduction. (R. E. Cramer—University of Illinois).

network of grain boundaries in which the equiaxed grains are considered to be true single crystals. Tammann (5) states for example: "Inside the polyhedral pattern there are evidences of a spharolite but the entire grain is a crystal." This is not necessarily true, since deep etch figures and X-ray diffraction studies do not support this view-

point. Elmore and McKeehan (6) have shown that ferrite grains have a "block structure." Therefore grains are not true single crystals, but are built of smaller units.

There are many reasons why grains should be divided into smaller units. It is easy to visualize that during the processes of recrystallization in the solid state, the grains interfere with one another's freedom of growth in all their selective directions, with the result that the grains are distorted. When the forces between grains are great enough, slip occurs on the glide planes at the heat treating temperature. In a previous paper (7) this was explained, proving that grains can actually become fragmented during heat treatment. A distorted grain or crystal was defined in the following way: "When a crystal or grain in its process of growth, assuming for the moment that it is growing freely in all its selective crystallographic directions,—encounters either other grains not in the process of growth or grain fragments unsuitably oriented relative to it for absorption, external forces resisting its growth are set up." This conception is supported by X-ray evidence and deep etch figures. This type of structure for instance is often developed during heat treatment of cold-worked metals. Zwicky (8) from theoretical considerations points out that when the rates of grain growth are great enough, plastic deformation should occur. This is in agreement with the experimental work of C. Nusbaum and the writer.

The conception of a "perfect crystal" is the ideal case rarely found in nature; however, some crystals such as the diamond and calcite may, under ideal conditions of growth, attain near perfection, nevertheless, they may have inherent "block structure." Physicists (7a) are studying the variations in the internal structure by various methods in an attempt to classify them; and determine what effects they have on the physical properties. Crystals which are not perfect but contain various types of imperfections have been called "mosaics." (When the blocks are slightly displaced from each other, the structure is called mosaic.) The idea of imperfection in the crystal structure of so-called perfect crystals was first conceived by C. G. Darwin as a possible explanation of the abnormal reflection of the X-ray beam from the crystal faces of rock salt. He assumed the crystal to be divided into small "blocks" slightly displaced from each other (only a few seconds of arc) in the reflecting surface.

Fibrous Texture

Fig. 4a is an X-ray diagram of a large single grain of 2.5 per cent

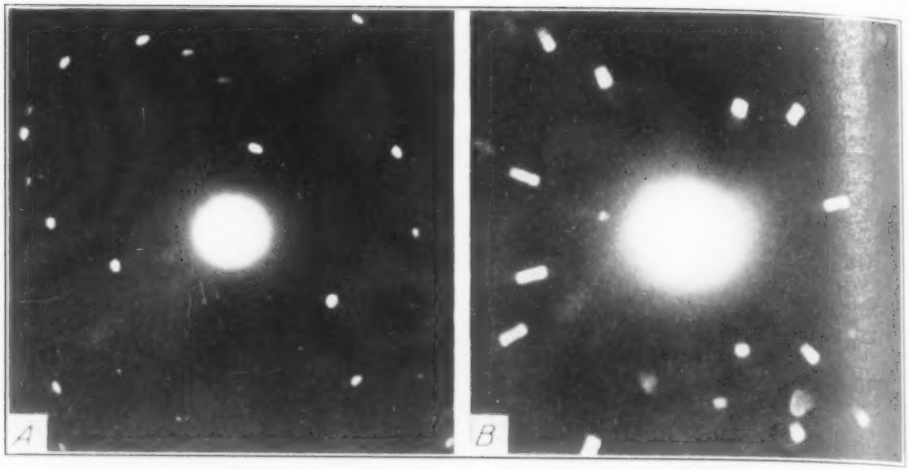


Fig. 4—X-Ray Laue Diagram of a Single Grain of Low Carbon Steel. A—X-Ray Beam Transmitted at Right Angles to the Surface. B—X-Ray Beam Transmitted at a Considerable Angle to the Surface but Through the Same Spot as in A.

silicon steel. It was mounted on a goniometer so that it could be set at any angle to the X-ray beam. The X-ray beam was transmitted at 90 degrees to the surface of the crystal. This X-ray diagram is characterized by a series of sharply defined diffraction spots, arranged on a definite geometrical pattern related to the orientation and symmetry of the crystal. On closer examination some of the diffraction spots are sharper than others, which means that some of the reflecting planes are not quite planar, indicating that even this structure is mosaic. This crystal was examined in 26 different positions over a wide range of angles and in only four positions were the X-ray patterns similar to Fig. 4a. In the others the single spots were replaced by radial streaks (elongated spots) (Fig. 4b) and some of these are resolved. The appearance of the radial streaks proves that the grain is sub-divided into blocks. These "blocks" are displaced relative to each other, being displaced more in some planes than in others.

The presence of "asterism"* when X-rayed in one direction and the lack of it in another can be explained if the grain is considered as built up of "blocks" which have one set of faces lying nearly in the reflecting plane, but in others the blocks deviate much more from the ideal position. The appearance of the radial asterism in heat treated metals or single grains is not at all surprising since it was already suggested by the deep etch figure., (Fig. 3) which clearly shows that the grains are composed of smaller units in which the "block" size ranges from 0.00015 to 0.0025 inch on the cube edge

*A.S.M. Handbook p. 84.

in silicon steel. The deep etch structure shown opposite page 80 of Desch's book (3) is even more impressive; here the rows of "blocks" are inclined at a considerable angle to each other.

It is preferable to call grain structures of heat treated metals which exhibit radial asterism when X-rayed, "fibrous textures" rather than mosaic, since the angular deviation of the block units is much larger.

The grain structure, however, is more perfectly developed in some cases than in others. For example, the grains in thin strip steels can be made quite perfect in structure, with the degree of perfection decreasing as the thickness of the strip increases. The thickness of the strip alone is not the only consideration, for in some alloy steels, as in silicon strip steels, grains quite perfect in structure can be developed in relatively thick strips. Perfect grains are considered to have inherent block structure.

In X-ray studies of this kind, the specimens for X-ray examination must be prepared with extreme care. The slightest amount of plastic deformation changes the structure, which the X-ray can easily detect. To avoid any possibility of deforming the specimen by plastic flow, the specimen is etched to the proper size in a 10 per cent nital solution. The specimen is usually made ½-inch square and after preparation it is mounted on the pinhole system of the X-ray spectrograph with the X-ray beam passing through the center. With these precautions taken into consideration, the X-ray diffraction pattern will be representative of the structure. In the next group of X-ray diffraction patterns, typical spectra of heat treated metals are shown, and prepared as described.

The target images in the X-ray diagram (Fig. 5a) are very sharp, even in the higher orders. This means that the grain surfaces reflecting the X-ray beam behave like optically flat surfaces. If these optically flat surfaces are composed of "blocks" slightly displaced, then the diffraction spots appear diffused (Fig. 5b). The X-ray method is so sensitive that even the slightest irregularity in the crystal structure can be detected. When Fig. 5a and 5b are compared, it can be seen that the diffraction spots on the outer rings of the former are very sharp, whereas in the latter they are very diffuse, with the exception of a few. This diffusion of the spots on the outer rings, is explained if the building blocks are slightly displaced relative to each other in each grain. The amount of displacement cannot be measured from the X-ray diagram, but the X-ray shows the existence

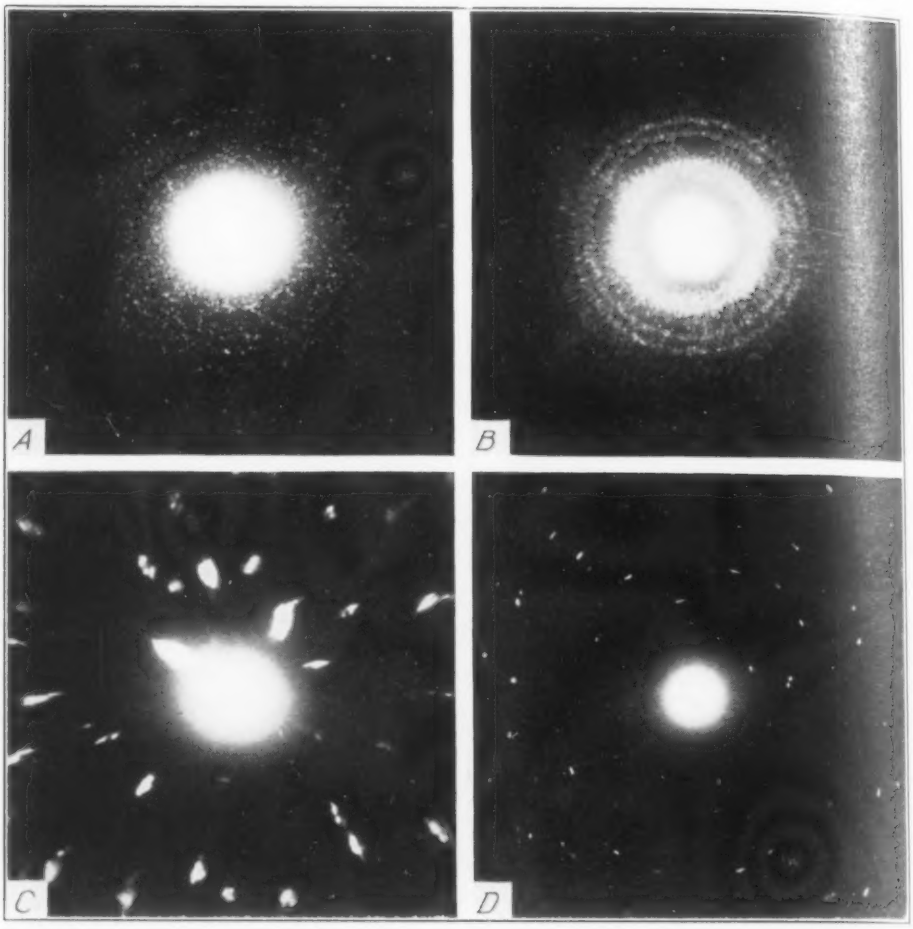


Fig. 5—X-Ray Diffraction Pattern Made by the Laue Method. A—Perfect Grain Having Inherent Block Structure. B—Exhibiting Mosaic Structure. C—Complex Fibrous Texture. D—Perfect Crystals (Having Inherent Block Structure).

of imperfect crystal structure. The mosaic structure was not developed by mechanical means, as by plastic deformation, but during heat treatment. Upon closer examination it is interesting to find one or two sharply defined spots on the outer diffraction rings (lost in reproduction but easily seen on negative). This means that some of the grains tend to be perfect in structure, even though they were grown under the same conditions as the distorted grains.

In the past a great deal of attention was focussed on the possibilities of perfect crystal structures. It was believed that if perfect crystal structures could be developed, then metals of unusual physical properties would result. X-ray diffraction studies do not support this viewpoint. Grains in metals have an "inherent block" structure and it is this block-like structure which controls the physical

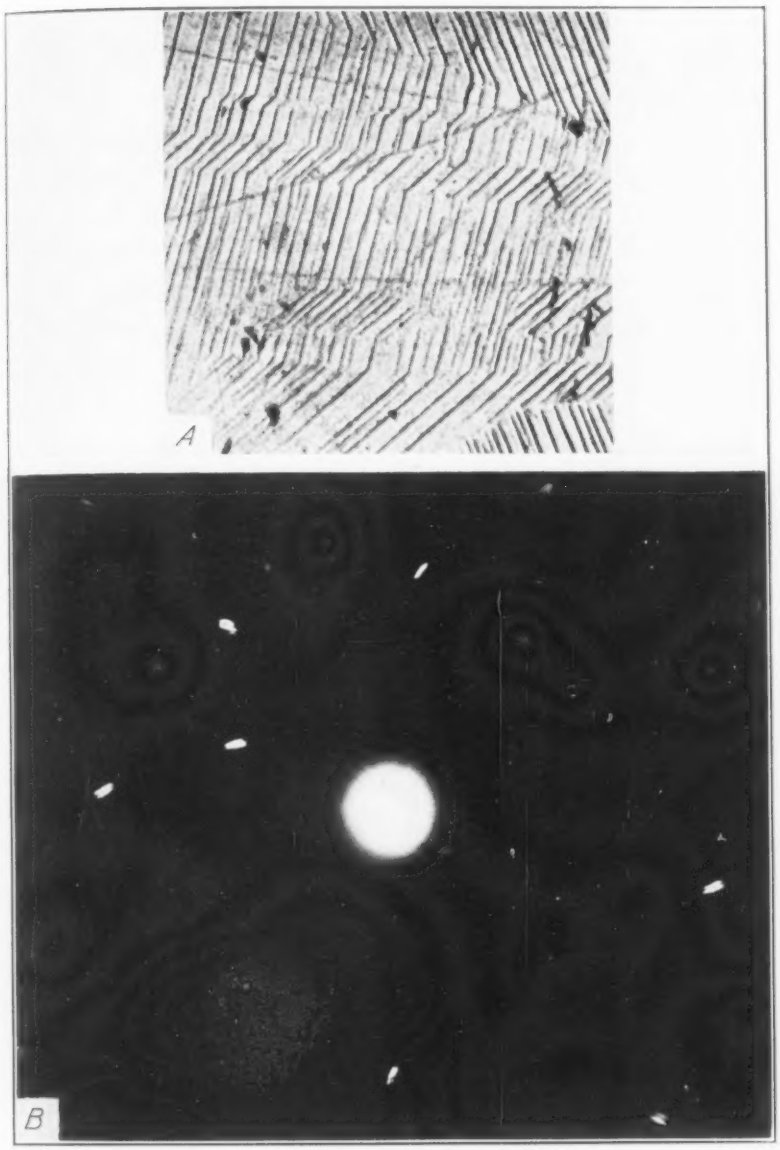


Fig. 6A—Slip Bands in a Single Crystal Caused by a Small Plastic Deformation, as Seen Under the Microscope.
Fig. 6B—A Single Crystal Slightly Deformed by Hammering. Note Splitting and Elongation of the Laue Spots. Compare with Figure 4A, Here the Spots are Elliptical.

properties. Crystals which have X-ray diffraction patterns indicating perfection, have an inherent block-like structure. This, as previously mentioned, was clearly demonstrated by taking a specimen and deforming it slightly. Examination under the microscope (Fig. 6) shows that the slip lines are equally spaced many atom planes apart, which is good evidence and supports the idea that grains are "block-like".

The X-ray diffraction spots of Fig. 5d appear sharp, but upon closer examination irregularities can be seen. These irregularities are attributed to certain kinds of imperfections in the grain structure. However in the light of our present knowledge of internal structure, we do not know just what effect these irregularities have on the physical properties. In Fig. 5c an X-ray diffraction pattern of a single iron crystal having imperfect crystal structure is shown. This is typical of a "fibrous texture" instead of mosaic, since it closely resembles the fiber diagrams of cold-rolled single crystals. The unit building blocks being displaced several degrees from each other. This X-ray diagram is representative of a complex structure, as is evident when the structure of the various diffraction spots are critically examined. These spots are resolved and diffuse, which means that the large blocks are further divided into smaller units, slightly and gradually displaced from each other (*7*). In some X-ray diagrams of heat treated metals the radial streaks are continuous which can be interpreted to mean that the blocks are small, and deviate gradually from each other but the total variation may be 1 or 2 degrees of arc. A viewpoint held by many authorities is that in annealed metals, diffuse diffraction spots and radial asterism is attributed to elastic lattice strains existing in the metal.

Relationship Between X-ray Structures and Physical Properties

The next group of X-ray diagrams show how the crystal structure is related to the physical properties of a 0.23 per cent carbon steel (Fig. 8). These X-ray diagrams were all made in the (*N*) direction (Fig. 7). The X-ray pattern of the cold-rolled strip (Fig. 7a) shows that the grains have been fragmented into smaller units by cold working, the crystal lattice has not been altered (except possibly for slight lattice displacements, small in comparison to the lattice dimensions). The increase in yield point (Fig. 8) and tensile strength and hardness, can be mainly attributed to the grain refinement imparted to this steel by mechanical working. Heating this strip at 900 degrees Fahr. for several hours and slowly cooling did not change the hardness or the tensile strength, but the yield point dropped slightly. The maximum softening of this strip steel occurred at 1300 to 1400 degrees Fahr. The X-ray diffraction patterns are consistent with these changes (Fig. 7b). The diffraction rings which

were uniform in texture, now consist of many spots. The grain fragments of the cold-rolled strip have recrystallized and the X-ray diagrams in addition show that they are not perfect crystals, but are mosaic and fibrous in texture. This strip steel was 0.040 inch thick and the X-ray specimens were carefully prepared by etching the large specimens in various acids, which assured uniform etching.

The curve (Fig. 8) also shows that as the heat treating temperature was gradually raised beyond 1500 degrees Fahr. the tensile strength, yield point and hardness also increased slightly. In higher carbon steels this increase is even greater.

The X-ray diagram (Fig. 7c) of the specimen heat treated at 1630 degrees Fahr. shows evidence of grain refinement with increased grain distortion (7). This specimen was cooled in the same way as the one heat treated at 1540 degrees Fahr. and the change in grain structure is striking, yet microscopic examination showed the grain size to be about the same for both specimens.

The Meaning of "Asterism" Found on X-ray Diagrams of Cold-worked Metals.

Many authorities on X-ray crystal structure analysis believe that "asterism" found on X-ray diagrams of cold-rolled polycrystalline metals, is an indication of a gradual change in the lattice parameters and referred to as lattice distortion. This viewpoint was presented by Czochralski (9) in 1923. He believed that the asterism indicated a gradual alteration in the spacing of the lattice planes.

G. L. Clark (10) also aheres to this interpretation of "asterism"; to quote, "Any pattern even for polycrystalline materials which shows these radial streaks is an indication of internal strain."

Other investigators do not agree with the interpretation offered by Czochralski and Clark. As a matter of fact, Ettich, Polyani and Weisenberg in 1921 studied the X-ray diagrams of severely worked metals, and concluded that the lattice parameters remain intact, (except for very small displacements due to stresses below the elastic limit). Later, Mark, Polyani and Schmid (10-A) studied the behavior of single zinc crystals deformed by extension and proved that the crystal lattice is unaltered by the deformation but the reflections of the X-ray beam from the reflecting planes became less well defined. A recent paper (11) claims that grain size also

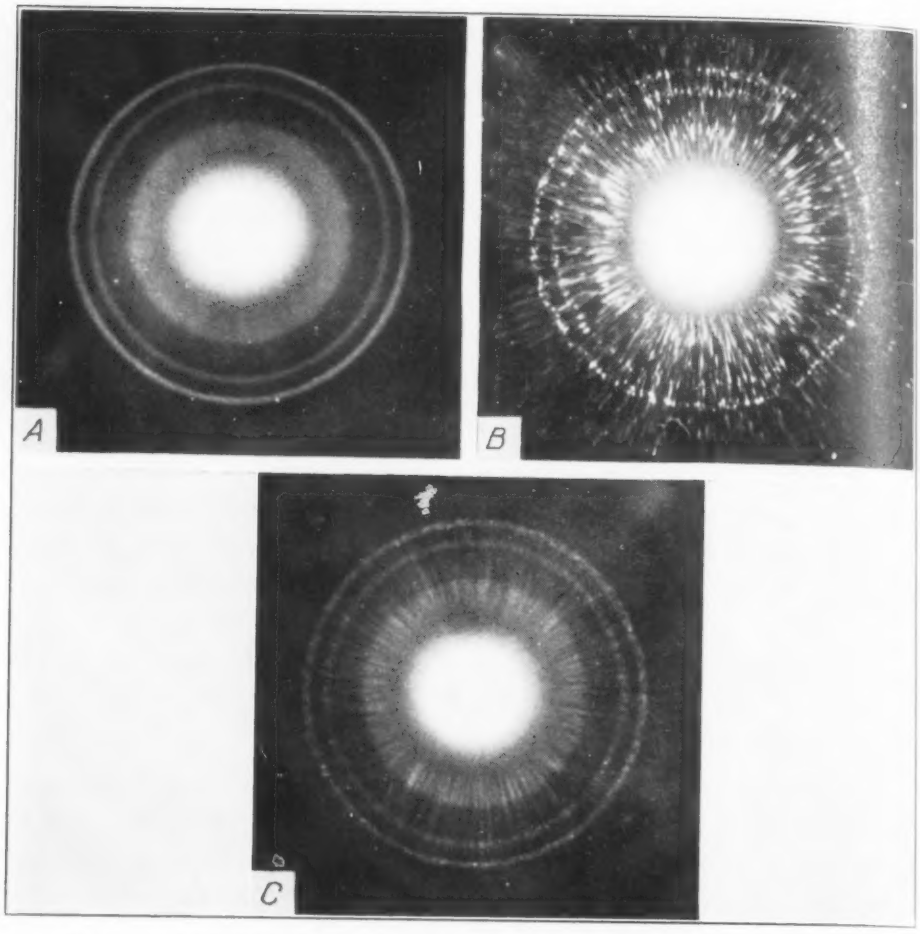


Fig. 7A—X-Ray Diagram of a 0.23 Per Cent Carbon Steel in the Cold-Rolled Condition.
Fig. 7B—After Recrystallization at 1300 to 1400 Degrees Fahr.
Fig. 7C—Cold-Worked Specimen After Heat Treatment at 1630 Degrees Fahr.

modifies the lattice dimensions. The situation is made even more complicated.

The point the author wishes to make is that radial asterism does not indicate the existence of lattice distortion, but shows that cold working deforms the grains into smaller units, or "blocks" which deviate gradually from each other. Lattice strains can and do exist but "radial asterism" does not indicate this. To detect the presence of elastic strains requires extreme precision in X-ray technique because these strains are small, and their exact measurement is made difficult because of such factors as "Poisson's ratio," and nonuniform strains throughout the grain structure.

Radial asterism in X-ray diagrams of metals given small area

reductions, indicate what has happened to the grains during plastic deformation. The grains are deformed into smaller block systems, which deviate slightly from block to block, and become smaller and smaller as the cold work is continued. This naturally influences the physical properties of the metal.

Furthermore, X-ray diffraction diagrams made of thin annealed strip steels stretched elastically, did not introduce asterism or diffuseness of the diffraction spots; this is good proof that "asterism" is not related to lattice strains (12).

The "block" structure of all heat treated metals offers a plausible explanation of the observed phenomena of selective slip planes equally spaced (see Fig. 6) and many atom planes apart, instead of having slip take place on every possible slip plane simultaneously. (Resolution of the radial streaks accounts for this structure in metals slightly deformed.)

Further proof that "asterism" does not mean the existence of internal strain can be illustrated by the following experiment. Cold-worked strip steel will not recrystallize when heated at temperatures below 950 degrees Fahr. If a cold-rolled single grain is heat treated at 700 degrees Fahr. the internal stresses should be relieved, and the elastically displaced atoms should return to their normal positions. The X-ray diffraction diagrams of deformed single grains before and after heating at 700 degrees Fahr. for 96 hours did not remove the "radial asterism." The same experiment was tried on a large-grained specimen of pure iron, tempered at 900 degrees Fahr. with the same result. This has also been confirmed by other investigators. (12a).

Plastic Deformation of Single Crystals and Polycrystalline Aggregates

Single crystals and polycrystalline aggregates are mosaic (13), fibrous in texture or when they appear perfect they have an "inherent block" structure which is easily detected when such structures are slightly deformed.

The question now arises: What influence does this "block" structure have on the mechanics of plastic flow? When a metal is gradually deformed, slip does not take place on every possible glide plane, but only on ones periodically spaced, many atom diameters apart. With continued cold work, more slip planes are created.

Ewing and Rosenhain (14) were the first to observe that the glide planes are periodically spaced when a metal block having a polished surface is slightly deformed. If the etched surface of a pure metal is given a small plastic deformation, the surface of the grains which appeared structureless before deformation, now is striated, and these are nearly equally spaced and have a step-like appearance. This mechanism of plastic deformation in metallic aggregates or in single crystals suggests a "block" displacement. This type of structure offers a reasonable explanation of the observed experimental fact, that all possible slip planes are not glide surfaces. We may, therefore, consider a crystal as divided into blocks periodically partitioned into glide planes of varying shear strength.

The "block" structure of grains is related to the "stretcher" strains found in low carbon strip steels given small plastic deformations by stretching. Roller leveling or pinch passing destroys this structure, since this operation produces more slip planes of uniform shear strength in the "blocks." Another way to look at it is that a finer block-like structure has been created. Metals which do not develop "stretcher" strains when slightly elongated, must inherently have a finer "block" structure in the grains to start with, or such grains have more planes of uniform shear strength to start with. The mechanism of plastic flow is more complicated in the case of "Luder lines," since the deformed areas are uniform, and spread over many grains in regions of uniform flow. This suggests some sort of directional weakness from grain to grain in the regions of uniform deformation. The correlation of "Luder lines" to the inherent "block" structures in the grain is a logical deduction, consistent with the mechanism of deformation, in that glide does not occur on every possible slip plane, but on glide "blocks" which in some metals are "large" and in others much "smaller," accounting for the presence and absence of "Luder lines" or stretcher strains. This explanation can also account for the drop of the beam at the yield point in low carbon strip steels. When the stretching load is applied, the stresses may be considered as concentrated on the glide blocks with the result that slip takes place on a limited number of glide planes, which have the weakest shear strength in each block. Glide continues for some time on one group until further shear is impossible due to increased resistance to flow, and then begins on another group of glide planes. The changes which occur in the structure of the metal when it is plastically deformed will be described briefly.

They are:

1. Slip takes place on the glide planes, spaced many atom planes apart.

2. With continued cold work, the grain fragments are oriented in a definite direction.

3. As slip progresses shear resistance to further slipping on these planes increases.

4. The mechanism of plastic flow appears to be one of block displacement, by rotation and translation.

5. The "blocks" or grain fragments become smaller and smaller as the cold working is continued, but the blocks retain their identity even after the severest plastic deformation—(Pulsifer).

6. The "blocks" are fragmented internally into particles of varying dimensions, some being colloidal. X-ray and electron diffraction studies prove that the lattice dimensions as well as the cubical symmetry is changed slightly as the particle size within the displaced blocks becomes smaller and smaller.

7. The density of the metal is decreased due to an increase in spacing between the displaced blocks (X-ray method cannot detect this).

8. The particles within the blocks tend to roughen the glide surface, and resistance to further slipping is increased, with the result that the metal hardens and its ductility is lowered.

9. Tempering causes a rearrangement of the particles within the blocks, however, the displaced blocks are not affected (asterism is not altered) but the $K\alpha$ doublets sharpen.

10. The block structure of the cold-worked metal is destroyed during recrystallization but a new block structure is again created, as already defined.

This outline will be discussed in greater detail in the appendix. (See page 1012).

X-ray diffraction diagrams of all heat treated low carbon strip steels are characterized by spotty Laue diagrams, in which the spots are more or less sharply defined. The sharpness or the lack of it proves that the grains are either perfect or mosaic in structure. However one must bear in mind that even in X-ray diagrams of metals having sharp diffraction spots the grain structure is inherently imperfect, and the X-ray method proves this. (See Fig. 6b).

X-ray Laue diagrams consisting of individual Laue spots, indicates that the grains are quite large, such grain structure is usually

soft and ductile. (Except in metals hardened by critical dispersion.)

For the purpose of illustration a low carbon strip steel was studied with the X-ray and the hardness value of each specimen was measured. The X-ray diffraction patterns of the heat treated specimens were always characterized by a spotty diagram. Similar to the one shown in Fig. (7B). The hardness value of the specimens was about Rockwell B-50. The X-ray Laue diagrams proved the structures to be mosaic and fibrous in texture. The spots on the innermost rings were quite easily recognized as individual spots but they were diffuse.

Cold rolling these strip steel specimens only 15 per cent, completely obliterated the identity of the individual diffraction spots, and the hardness increased from Rockwell B-50 to Rockwell B-75. The grains have been severely fragmented into smaller units by only an area reduction of 15 per cent. (The X-ray diagrams of such specimens show the grains to be fragmented; but as a rule they do not lose their identity when given only small area reductions.) Microscopic examination of such specimens would only indicate a mild fragmentation. The X-ray method clearly shows that the fragmentation has been severe, and is in line with the marked increase in hardness.

Extensive experiments prove that the strength of metals is influenced by the size of the grains, and fractures invariably take place through the crystal and not at the grain boundary. (At room temperature.) This would mean that the grain boundary strength exceeds the crystal strength. These facts have been repeatedly proven by research work on single crystals and crystal aggregates in which the grain size varied over a wide range. In the polycrystalline material the obstruction to flow caused by the surrounding grains having unsuitable orientations builds up resistance at the grain boundary. In the single crystal, gliding on the slip planes is not hampered in this way and so extension continues.

The curve of Fig. 9 compares the elongation of a small grained zinc strip and a single crystal of the same material. The single crystal could be elongated 400 per cent and a breaking load of only 1.8 kilograms per square millimeter (2560 pounds per square inch) was required. In the polycrystalline material the strip failed after elongating only 40 per cent, but required a breaking load of 15 kilograms per square millimeter (21,330 pounds per square inch). The resistance to "block" gliding at the boundary reduced the elongation and increased the breaking load 8.5 times. Another fact proven in

these experiments on zinc crystals is that glide does not take place on every possible slip plane. The glide "blocks" are inherently present in the grains even before extension begins. In other words, the shear strength varies from point to point within the "block," with the result that plastic flow does not begin on all of the slip planes at the same time. In single crystals elongated by stretching, slip

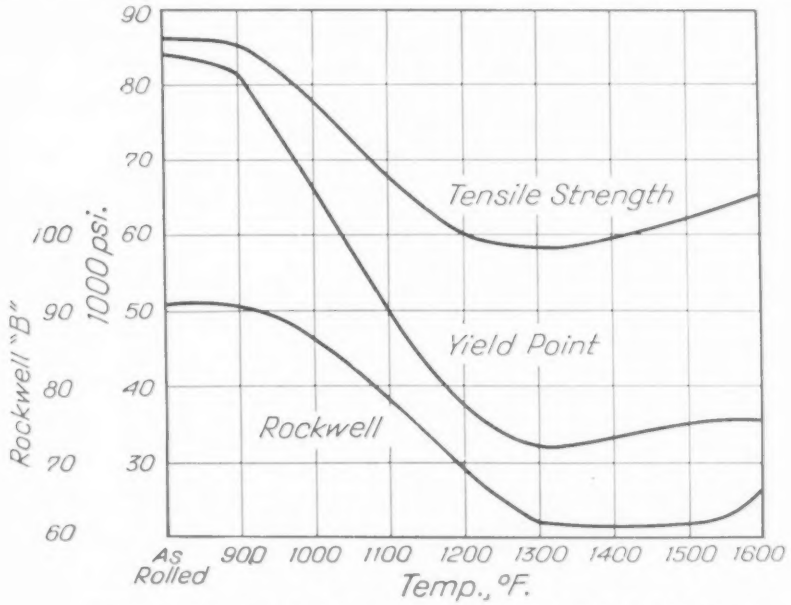


Fig. 8—The Physical Properties of the Cold-Rolled 0.23 Per Cent Carbon Steel After Heat Treatment at Various Temperatures.

often continues on the original slip planes until failure. Numerous markings appear on the "glide blocks" (15) indicating that other "glide surfaces" did tend to slip, but they had greater shear strength.

In cold rolling single crystals, new glide planes are continually being created with the result that single crystals can be deformed almost without limit, the only limitation being the mechanical limitations of the cold rolling apparatus.

Simple glide is said to occur in crystals of the zinc type since gliding is known only to occur on the 0001 planes. In metals having the cubic type of space lattice, gliding occurs on systems of glide surfaces. The phenomenon of gliding on slip planes during plastic deformation has never been explained from an atomistic point of view, in spite of the fact that we know that gliding occurs on atomic planes. Our conception is entirely a mechanical one, but the X-ray has guided us in the determination of the glide planes and also the direction in which the grain fragments are oriented when a metal is

plastically deformed. The microscope and the X-ray both show that plastic flow is a block phenomenon in which the "glide blocks" are nearly equally spaced, and gliding can only occur when work is done against the lattice forces which resist extension by shear.

When a metal is plastically deformed by cold rolling or by wire drawing, the grains are fragmented and oriented with certain crystallographic directions parallel to the direction of working. In

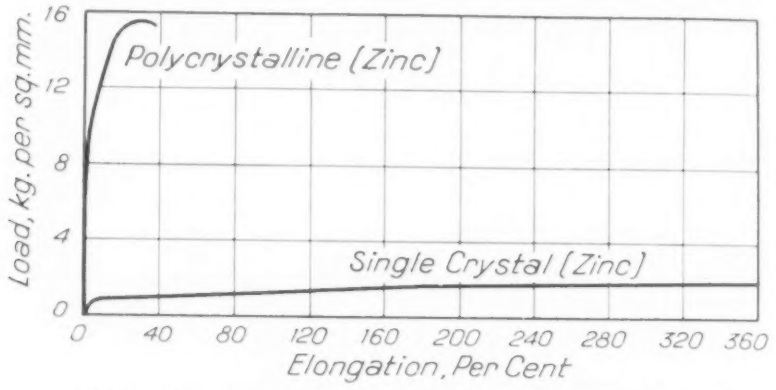


Fig. 9—The Physical Properties of Fine-Grained Zinc Strip Compared with a Single Crystal of the Same Material.

alpha iron, it is now well established that after severe reductions by cold rolling the [1 1 0] direction of the grain fragments tends to coincide with the rolling direction; however, there is a lack of agreement concerning the planes on which slip takes place during plastic deformation, and which planes lie in the rolling surfaces. A number of papers have appeared in recent years which show that slip is possible on the {1 2 3} and the {2 1 1} planes. Gough's (16) investigations on single crystals support these results.

Quite recently several other investigators (17) believe that they have observed slip also to occur on the {1 0 0} planes in single iron crystals. The possibility of slip on the {1 0 0} planes is plausible in the case of single crystals but not so likely in the case of polycrystalline iron aggregates, especially of the commercial variety. However, the {1 0 0} planes have been considered to be the cleavage planes in some alloys of iron, which are brittle.

Past experience on crystal aggregates of iron and iron alloys deformed by cold rolling proves that slip is most likely to occur on the {1 1 0} planes. These planes have the greatest atomic concentration and spacing and should, from dynamical considerations, have the weakest atomic bonding and therefore the greatest susceptibility to slip. Apparently under certain conditions slip appears possible on

the {2 1 1} planes (18) but it may also mean that the {2 1 1} planes tend to lie in this direction with slip occurring on the {1 1 0} planes.

Another fact which seems to be strong proof that slip must be experienced on a limited number of crystal planes (1 1 0), (1 1 2), etc., in alpha iron lies in the similarity of the X-ray fiber diagrams. The patterns would suggest that slip is most likely to occur on the {1 1 0} planes in polycrystalline aggregates of alpha iron. In tensile test* strips of low carbon strip steels elongated to failure slip occurred on the {1 1 0} planes.

Perhaps the explanation of the manner in which a single crystal of alpha iron deforms cannot be applied to a polycrystalline aggregate without restrictions.

Fig. 10 is a group of fiber diagrams made of three neighboring grains after an 86 per cent reduction in thickness. The direction of plastic flow in each of these grains was in the same direction (approximately). The marked difference in the appearance of these X-ray diagrams is due to different crystal planes, lying in the surface of the strip, but their indices were not determined.

In small-grained iron samples, drastically cold-worked, the X-ray diagrams are all similar in appearance in all parts of the strip. Fig. 11, is a typical X-ray diffraction pattern made of a low carbon strip steel (0.08 per cent) given an area reduction of 97½ per cent by cold rolling. This area reduction has oriented the grain fragments with the [1 1 0] in the direction of rolling. These X-ray diagrams, Figs. 10 and 11, have one thing in common, the [1 1 0] direction tends to lie in the direction of rolling; however, the deformed single crystal diagrams, as already pointed out, suggest that different planes lie in the surface of sheet. If the same plane lies in the surface of the large-grained strip, identical X-ray fiber diagrams would result (see Fig. 10).

This problem was considered in the case of various alpha iron alloys and low carbon steels, with the result that in low carbon steels the {1 1 0} plane was most prominent in the rolling surface, but the {2 1 1} planes were also strong. The {3 1 0} {3 1 1} {1 1 1} {3 2 1} planes were also definitely present, but very weakly so.

From these experimental researches, it is possible to arrive at a definite conclusion; that the planes which lie in the surface of the cold-rolled metals depends upon the chemical composition. Several planes can lie in the surface of the sheet, with one or two planes

*To be shown in another portion of this paper.

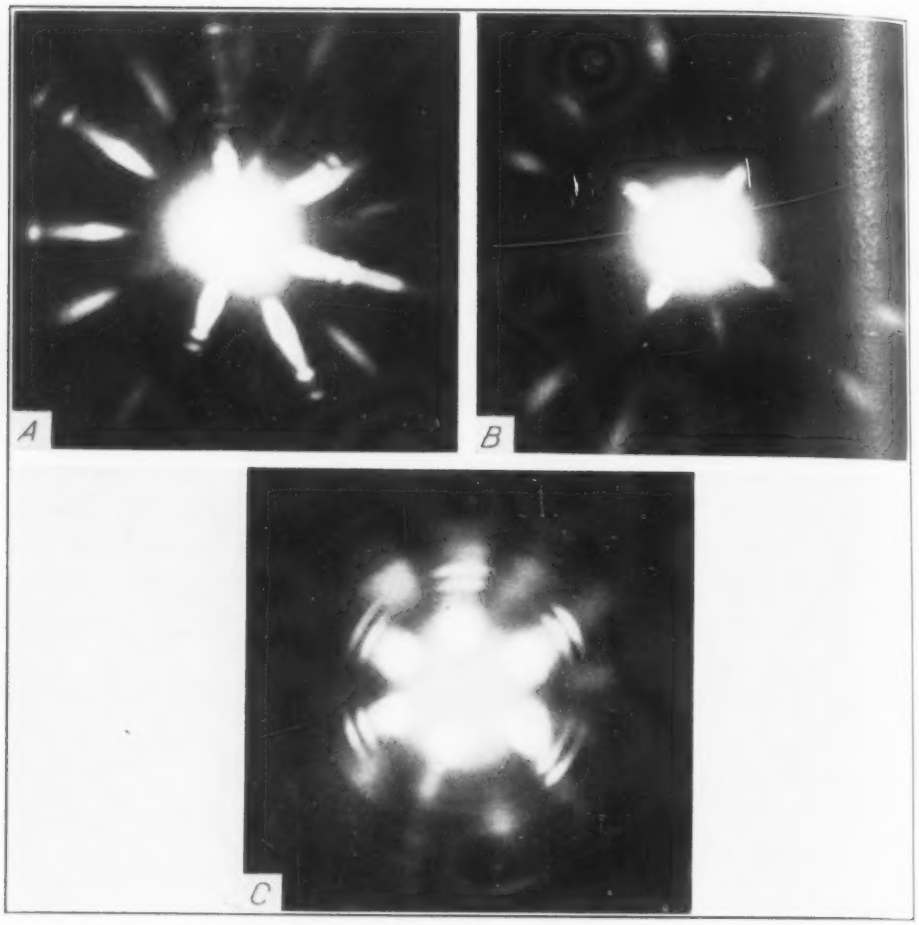


Fig. 10—X-Ray Laue Diagrams of Three Neighboring Single Crystals After an Area Reduction of 86 Per Cent in Thickness. These Diagrams all Show that the [1 1 0] Direction Tends to Lie in the Direction of Rolling but Different Crystal Faces Lie in the Plane of Rolling.

predominating, followed by others not as strongly present. The distribution of the grain fragments in the surface of a cold-rolled sheet has a marked influence on the physical properties, with and cross grain. The distribution just described is not a symmetrical one, about the rolling direction; and accounts in part for the variation in the tensile strength, with and cross grain. The older theory of grain distribution in cold-rolled strip or sheet assumed a symmetrical arrangement, but this could not explain the great difference in the physical properties with and cross grain.

Preferred Orientation in Hot-Rolled Strip

Very little is reported in the literature about the structure of

hot-rolled metals and alloys. However, cold-rolled metals have been studied in greater detail.

It is of considerable importance to determine where hot rolling ends and cold rolling begins, as the temperature of the metal is progressively lowered during the mechanical working. The dividing

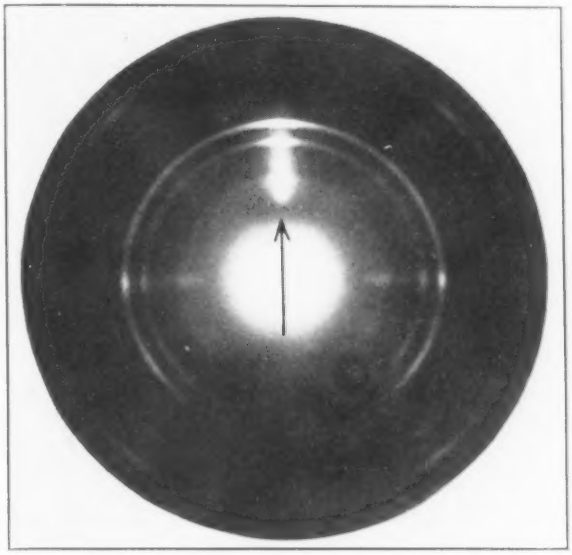


Fig. 11—X-Ray Diffraction Diagram of a Low Carbon Strip Steel Severely Cold-Worked. Arrow Indicates Direction of Rolling.

line has been arbitrarily defined as the temperature at which the metal fails to recrystallize with continued mechanical working.

In many hot-rolled alloy strip steels, a fiber structure is developed during hot rolling, which is difficult to remove by heat treatment. Hot-rolled low carbon strip steels may or may not retain a fiber structure after hot rolling; this depends only on the amount of rolling below the A_3 point. When finished above A_3 the structure recrystallizes on passing through the A_3 and the grains are usually oriented at random. If the finishing passes are too severe, then directional properties can be retained even though the strip is finished above the A_3 (slightly). Finishing a low carbon strip steel near the A_1 point always produces a grain structure having preferred orientation. A Bessemer strip steel is more apt to retain directional properties than an open-hearth steel of the same chemical analysis. The slag stringers and other inclusions in the Bessemer steels favor the retention of the directional properties.

A few X-ray diffraction patterns will aid in proving the existence of preferred orientation in hot-rolled strip, microscopic examination does not show this, so the X-ray method is unique in this respect.

The hot-rolled strips examined were X-rayed as shown in Fig. 12. The letters (P), (N) and (T) refer to the directions in which the X-ray beam was transmitted through the specimen. To save time

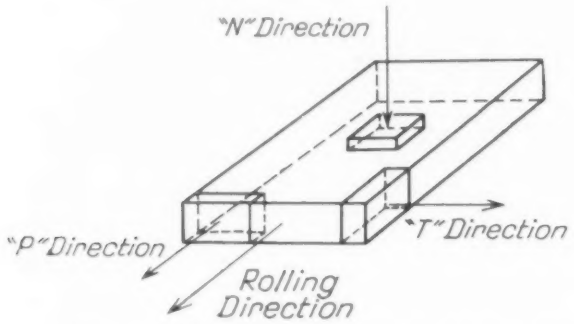


Fig. 12—Schematic Representation Showing the Manner in Which the Specimens Were X-Rayed in Hot-Rolled and Cold-Rolled Strips.

they will be referred to as (P), (N) and (T) with the understanding that they were prepared and X-rayed as shown in Fig. 12.

A strip of low carbon steel 0.109 inch thick and finished above the A_3 point on a continuous strip mill was examined in this way. It was free of (Fig. 13) directional properties, as should be expected, since it was finished above the A_3 and the reductions per pass were not excessive on the finishing stands. These two factors are important when the elimination of directional properties is required. Experiments prove the directional properties most intensely developed in the (T) direction in all hot-rolled metals. In some instances strong fiber diagrams were also found in the (P) direction. Fig. 14 was made of a hot-rolled 1 per cent silicon steel strip X-rayed in the (T) direction; analysis shows that the [1 1 0] direction coincides with the rolling direction, and the {2 1 1} and the {1 0 0} planes predominate in the plane of rolling. Other crystal planes definitely but only weakly present in varying proportions were the {1 1 1} {1 2 3} {3 1 0} planes, etc.

Regardless of the distribution of the crystal planes which lie in the surface of the hot-rolled strip (alpha iron alloys), it is apparent that the [1 1 0] direction tends to coincide with the rolling direction or lies in the plane containing the (P) and (N) directions.

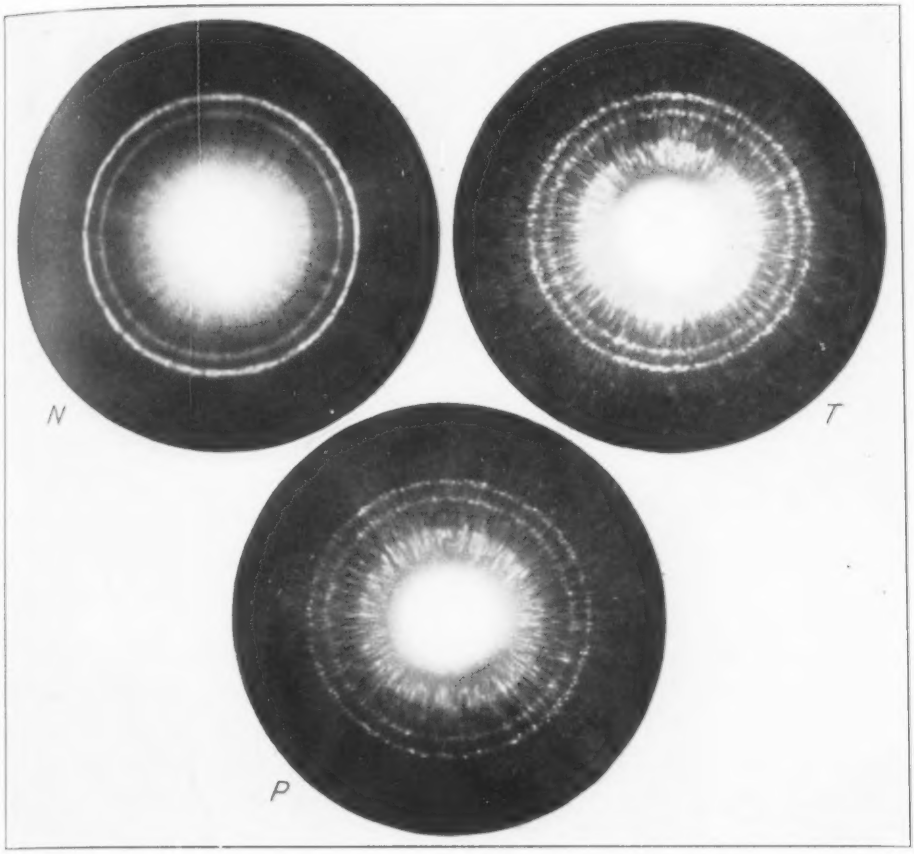


Fig. 13—X-Ray Diffraction Diagrams of a Low Carbon Hot-Rolled Strip, X-Rayed in the N, T and P Directions.

The grain fragments are not oriented ideally but deviate by some degrees from the ideal positions; in other words, the planes are distributed statistically about the ideal position. A number of things are responsible for this, a few of which are:

(1) Chemical composition
(2) Rolling conditions
    (a) Starting and finishing temperature
    (b) Reduction per pass
(3) Roll diameter

Fig. 14b is an X-ray diagram of a low carbon steel hot-rolled strip finished near 1500 degrees Fahr. and we see that strong fiber structures can also appear in the (N) direction, in addition the grain fragments have partially recrystallized. The most reasonable explanation for the appearance of this rather strong fiber pattern

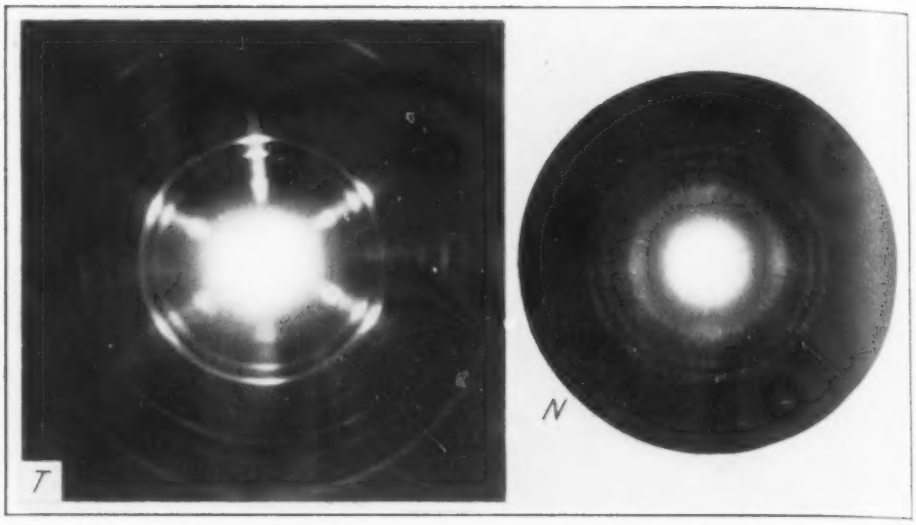


Fig. 14a Fig. 14b

Fig. 14—X-Ray Diffraction Diagrams of Hot-Rolled Strips, X-Rayed in the T and N Directions.

in the (N) direction is that the finishing passes made heavy reductions below A_3. These X-ray diagrams illustrate the value of the X-ray in studying hot-rolled structures. This method of analysis reveals precisely what changes occur in the internal structure of the grains during hot rolling. By controlling the hot rolling many types of structures can be developed; for example, the hot-rolled structure, Fig. 13, will not develop ears since the directional properties have been entirely eliminated, however, the structures shown in Fig. 14a and b have directional properties and such metals will have a tendency to tear in the die and form ears during deep drawing operations. Furthermore, when the diffraction spots lack sharpness of definition and are elongated radially (Fig. 13), then the grains are "block-like" structures in which the grains are either mosaic or "fibrous."

To further illustrate the complexity of the hot-rolled structures developed during hot rolling, another group of X-ray diffraction patterns typical of the 14 per cent chromium and 8 per cent nickel steel and nickel hot-rolled strips are shown. The specimens (Figs. 15 and 16) were X-rayed in the (P), (N) and (T) directions. See Fig. 12. The fiber structures in the (T) direction in the alloys are similar. The grain fragments in the 8 per cent nickel steel and the 14 per cent chromium steel both have the [1 1 0] direction in the direction of rolling with the {2 1 1} and the {1 0 0} planes predominating in the

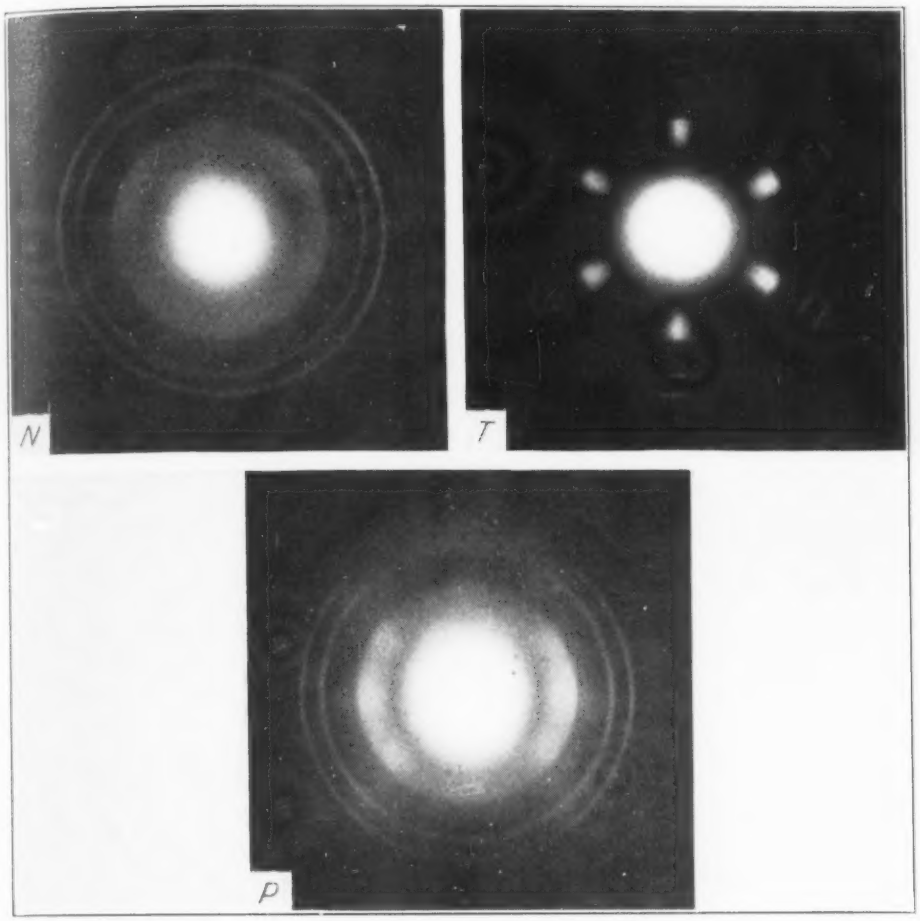


Fig. 15—X-Ray Diffraction Diagrams of a 14 Per Cent Chromium Steel Hot-Rolled Strip, X-Rayed in the P, N and T Directions.

plane of rolling. Both of these steels are body-centered cubic metals while nickel has a face-centered cubic space lattice.

In the hot-rolled nickel strip, see Fig. 17, the planes having the greatest atomic concentration are the (1 1 1) planes and are the most probable slip planes. The six innermost spots and rings are due to reflection of the X-ray beam from the (1 1 1) planes. From the arrangement of the diffraction spots relative to the direction of rolling, we find that the [1 0 0] direction and the [1 1 1] direction lie in the direction of rolling. The planes which lie in the plane of rolling were not determined. To find the planes which lie in the surface, the reflection method is best suited for this type of work. The specimen to be analyzed is placed in a reflection cassette and set at various predetermined angles to the X-ray beam so as to reflect from those

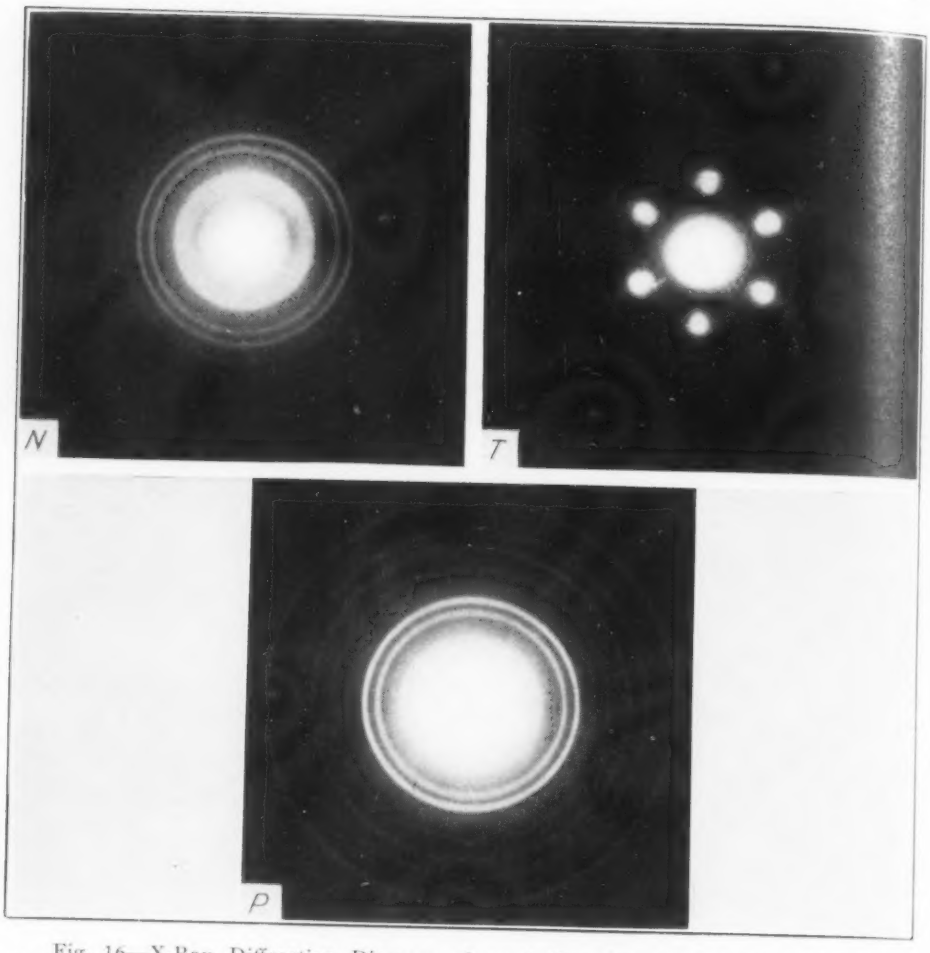


Fig. 16—X-Ray Diffraction Diagram of an 8 Per Cent Nickel Steel Hot-Rolled Strip, X-Rayed in the N, T and P Directions.

planes which might be expected to lie in the surface. When the specimen is set in these various positions, some plane will reflect strongly in some positions, weakly in others, and when no line appears then that plane is not present in the surface of the sheet. It should be pointed out that these planes do not all lie exactly in the plane of rolling, but are distributed about the ideal position in a statistical manner. To determine the limits of distribution an X-ray goniometer must be used. The statistical distribution of the grain fragments in a hot-rolled sheet is quite important. The (T) fiber diagrams of two different alpha iron alloys are similar, then the (P) and (N) diagrams would naturally be expected to appear the same. Comparing the corresponding X-ray diagrams, (T), (P) and (N) for the 8 per cent nickel steel and the 14 per cent chromium steel, we find that the (T) and (N) diagrams are similar but the (P) fiber

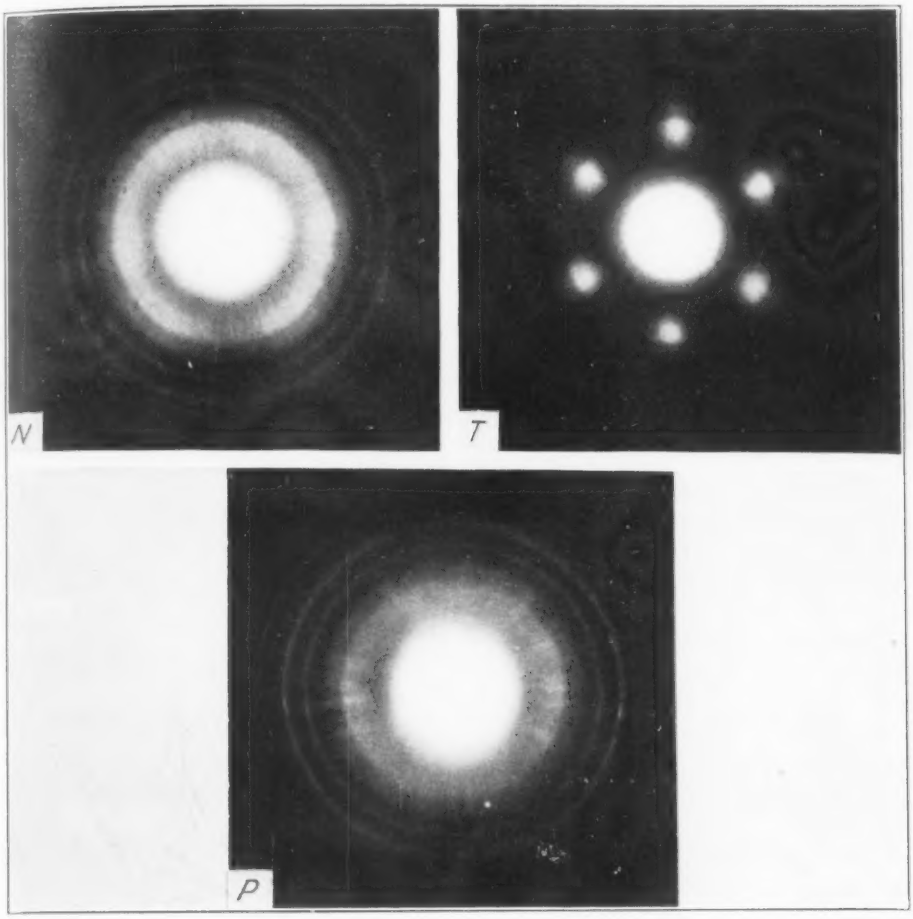


Fig. 17—X-Ray Diffraction Diagrams of a Hot-Rolled Nickel Strip, X-Rayed in the N, T and P Directions.

diagrams are entirely different again. The difference in the (P) and (T) diagrams account for the difference in the physical properties, with and cross grain or in any other direction in the surface of the strip. Also, since the (P) diagrams for these two alloys differ so much from each other and the (T) diagrams are the same, naturally the cross and with grain properties of these alloys must differ proportionately.

The difference in the (P) and (T) diagrams of the 14 per cent chromium steel and the 8 per cent nickel steel accounts nicely for the difference in the with grain and cross grain properties. If the (P) and (T) diagrams were alike, then certain physical properties in these two directions would be about equal. In the low carbon strip steel (Fig. 13) the (T) and (P) diagrams do not differ and the physical properties were found to be the same, which is good proof

that the fiber diagrams (variations) and physical properties in different directions are related. The distribution of the grain fragments or crystals therefore influences the physical properties in various directions. The X-ray is, therefore, indispensable in determining the variations found in the fiber structure, and the internal structure of the grains of hot-rolled strip. Directional properties are without question the major source of trouble in deep drawing and stamping operations. Where directional properties are present

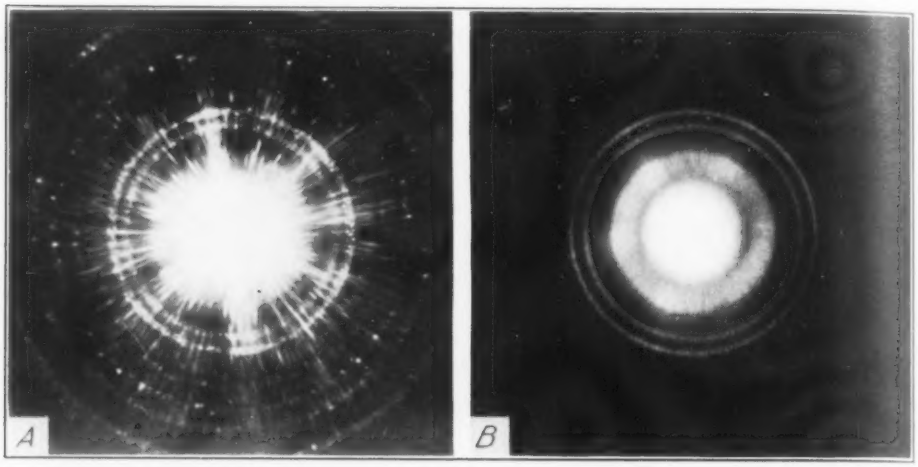


Fig. 18A and B—X-Ray Diffraction Diagrams of 14 Per Cent Chromium Hot-Rolled Strip Heat Treated at Various Conditions.

in alloy steel hot-rolled strips, it is sometimes impossible to remove them even by heat treatment. We know that the directional properties in low carbon strip steels are easily removed by simply finishing the strip above the A_3 point and finishing with not too severe reductions on the finishing stands. This is not always possible, in the 14 per cent chromium steels; partly because there is no A_3 transformation to help. The next alternative is to find out whether or not some form of heat treatment will remove it. The next group of X-ray diagrams (Fig. 18) were made of a 14 per cent chromium steel alloy all X-rayed in the (T) direction since the fiber structure is most intensely developed in this direction. A number of different heat treatments were tried, but without success. Some specimens were even quenched in oil and water from temperatures ranging from 1200 to 1800 degrees Fahr. and with the same result as before. Long time heat treatment apparently had no more effect on the directional properties than short time heat treatments. This meant

that the directional properties, if they are to be eliminated in this steel, must be eliminated by proper hot rolling.

Preferred Orientations in Hot-Rolled Rods

It is well known that when hot-rolled rods are drawn through a wire drawing die the grains are fragmented into smaller units and with continued drafting are oriented with the [1 1 0] direction in the direction of drawing, and distributed entirely at random about the axis of the wire. The X-ray diagrams (Fig. 19a) prove that the grain fragments are oriented in this way after plastic deformation through a wire drawing die.

The fiber structure in the case of the hot-rolled rods (which are usually rolled from 1¼ x 1¼ inch billets into rods of various gages) is not developed in this way. X-ray diffraction analysis proves that the fiber structure in the hot-rolled rods actually develops at right angles to the rod axis with the fiber axis parallel to the axis of the work rolls of the rod mill. Fig. 19b illustrates this type of structure. Comparing it with the corresponding fiber diagrams of the cold drawn wire, the differences in these structures can be readily appreciated by comparing the corresponding (P) and (N) directions. These experiments prove in a simple way that the fiber structure of the hot-rolled rod differs from the structure developed during the cold drawing operation.

The reason for the development of the "cross fiber" in the hot-rolled rods is not difficult to explain if one considers the manner in which the rod is rolled from the billet. The alternate passes are diamond and oval and the last pass is oval to round. The reduction along the major axis of the oval is very large and accounts for the peculiar fiber structure found in hot-rolled rods. The working of the rod from oval to round on the last pass may be considered as an upsetting action along the major axis. This accounts for the fiber structure developing in this direction. The fiber structure is not influenced by the type of hot mill, but depends merely upon the type of passes used on the rod mill.

A rod bundle was studied at various points throughout its entire length and it was surprising to find that the cross fiber was not uniformly developed throughout its entire extent, and the perfection of the grains also varied. The cross fiber, therefore, affords a sensitive method in the study of the uniformity of hot-rolled rods.

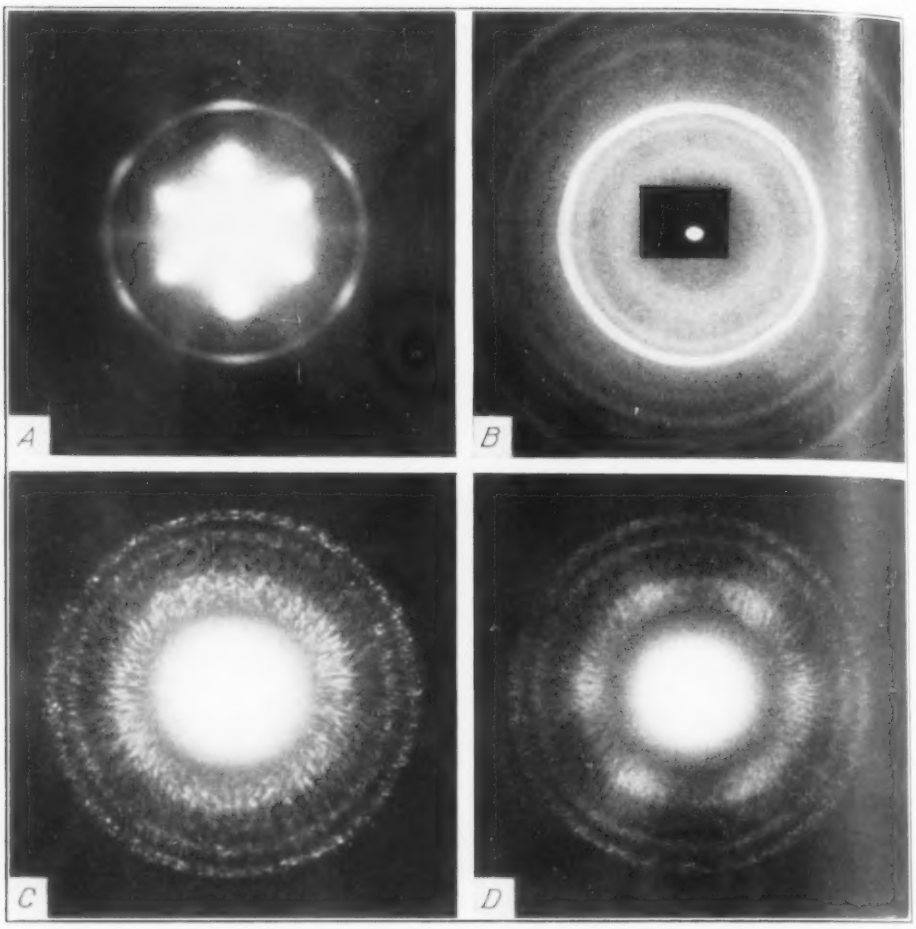


Fig. 19A—X-Ray Diffraction of a Hard Drawn Wire, X-Rayed at Right Angles to the Wire Axis.
Fig. 19B—X-Ray Diffraction of the Same Hard Drawn Wire but X-Rayed Parallel to the Wire Axis.
Fig. 19C—X-Ray Diffraction of a Hot-Rolled Rod, X-Rayed at Right Angles to the Rod Axis.
Fig. 19D—X-Ray Diffraction Diagram of Same Hot-Rolled Rod but X-Rayed Parallel to the Rod Axis.

Effect of Small and Large Plastic Deformations on the Physical Properties, and the Crystal Structures.

It is well known that plastic deformations by hot and cold rolling change the physical properties and the crystal structure. Cold rolling is more effective than hot working, in increasing the elastic limit, yield point, tensile strength and hardness, and at the same time decreasing the ductility and elongation. The increase is proportional to the amount of cold rolling. The change in hardness when a low carbon hot-rolled strip is cold-rolled is shown in Fig 20. The hard-

ness increases rapidly at first but with continued cold rolling the increments become less. As a rule a low carbon (0.05 per cent — 0.10 per cent carbon) hot-rolled strip cannot be made harder than Rockwell B-97 by cold rolling.

The X-ray has demonstrated beyond all doubt that when a metal is severely cold-worked, the lattice parameters are not changed appreciably, except for very small displacements which are due to elastic stresses and the metal is still crystalline.

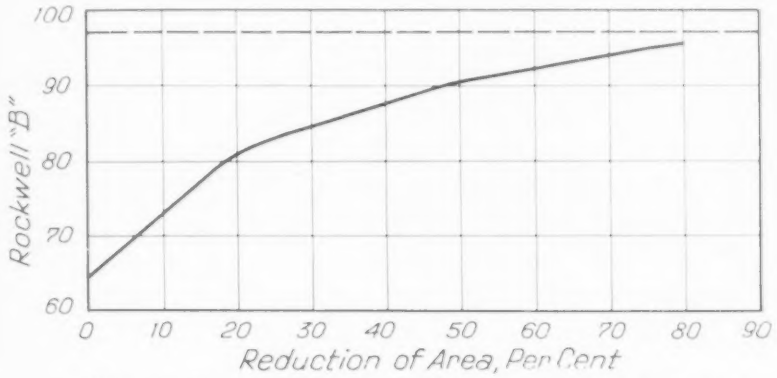


Fig. 20—Showing How Hardness of a Low Carbon Strip Steel Changes When Given Various Area Reductions.

Sometimes the lattice is transformed into another type, when cold-worked. This actually happens when 18-8 is drastically cold-rolled, the alpha phase is transformed from the gamma phase, and proves that 18-8 is metastable when plastically deformed. This lattice transformation can be easily detected with the X-ray, but the amount transformed is very small, never exceeding 5 per cent.

The X-ray is indispensable in studying the effect of small area reductions on the grain structure of polycrystalline metals as well as single grains. Small area reductions up to 15 per cent do not elongate the grains sufficiently for detection under the microscope. If this were the only method available then it would appear that small plastic deformations did not change the grain structure of the annealed metal appreciably. When an annealed low carbon steel wire is given an area reduction of only 5 per cent, the physical properties are changed considerably as shown in the table.

Low Carbon Steel Wire (0.06 Per Cent Carbon)

	Yield Point	Tensile Strength	Per Cent Elong. in 2 inches
Heat treated (0.0375" Dia.)	17,400	41,900	48%
Area Reduction, 5%	32,300	47,800	23%

This small reduction in area has practically doubled the yield strength, and lowered the elongation in about the same ratio, but the tensile strength was not changed very much. Figs. 21a and 21b show the X-ray structure before and after an area reduction of 5 per cent. A great change has occurred in the crystal structure, consistent with the changes in the physical properties. This small plastic deformation has fragmented the grains into smaller units. The outer diffraction rings of the undeformed sample consist of rather sharply defined diffraction spots, but these become diffuse in structure when this specimen is reduced 5 per cent in area (cross sectional); since the grains have been broken up into smaller units. This is clearly seen if these X-ray diagrams are compared the diffraction spots of the outer rings are now ill defined, showing that the grains have been fragmented into smaller units or blocks, purely by mechanical means. (The grains have been made "fibrous in texture" by cold rolling, due to rotation of the displaced blocks on the glide surfaces.)

An area reduction of 15 per cent by cold rolling is sufficient to completely fragment these grains (Fig. 21c) in the sense that the identity of the individual diffraction spots has been obliterated. Further cold rolling does not alter the structure of the diffraction rings, but continued cold rolling will eventually induce directional properties when the sheet is X-rayed perpendicular to the surface of the strip. (It was shown in another paper (19a) that the directional properties are imparted to strip steels when given small reductions, which can be detected when the X-ray beam is transmitted in the (T) direction (Fig. 12). Cold rolling has increased the yield point, ultimate strength and the hardness. These changes are attributed to the surprisingly large amount of grain fragmentation introduced by such small reductions in area. The lattice energy is increased by cold working and this decreases the density of the metal slightly, as already explained.

The grain size of metals is refined by mechanical working into many blocks slightly displaced relative to each other. The changes in the physical properties are accounted for by the grain refinement and change in the energy content of the block structure.

This interpretation is in line with the work of Mark, Polanyi, Schmid (19). They studied single crystals of zinc in the form of wires and elongated them by stretching, and among other things, concluded that the crystal lattice was unaltered by elongating the

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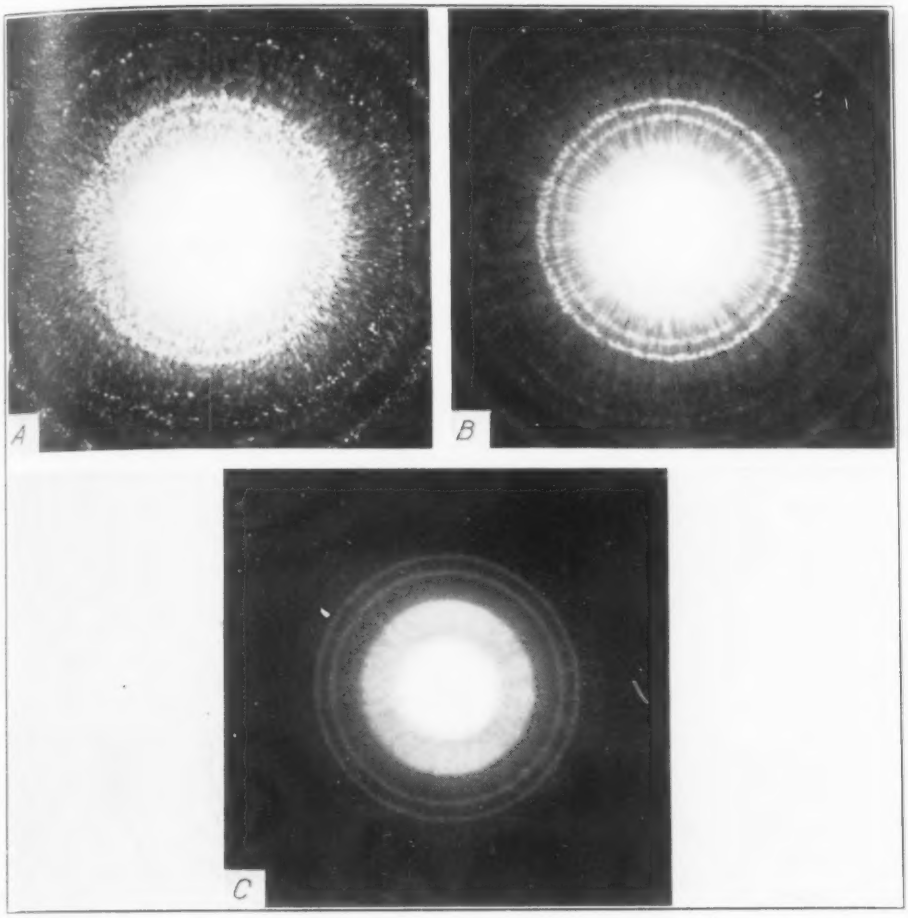

Fig. 21A—X-Ray Diffraction Diagram of a Normalized Cold-Rolled Strip Showing Mosaic Grain Structure.

Fig. 21B—X-Ray Diffraction Diagram After Having Received an Area Reduction of Only 5 Per Cent. Note the Tremendous Change in the Crystal Structure.

Fig. 21C—After Having Received Area Reduction of 15 Per Cent. Note that the Rings are Now Practically Uniform and Continuous in Character.

crystal, but the reflection of the X-ray beam from the crystal faces became diffused. These results can be applied to this problem in which polycrystalline aggregates were investigated instead of single crystals.

When a strip steel specimen is elongated in a tensile testing machine, the grains are fragmented into smaller units or blocks by translation and rotation and with continued loading of the tensile specimen, the grain fragments are gradually oriented with the [1 1 0] coinciding with the axis of pull. If the loading is carried far enough rupture occurs. The failure is due to shear on the (1 1 0) planes (Fig. 22) and these are inclined at 60 degrees to the axis of tension.

One might believe that shear failure should occur at 45 degrees to the axis of tension as demanded by the maximum shear theory. This is impossible for the following reasons: The grains at the time of failure are oriented with the [1 1 0] directions parallel to the axis of tension. Calculation shows that the other possible [1 1 0] directions make angles of 60 and 90 degrees to the one which lies in the direction of rolling. Shear cannot take place on the plane which lies at right angles to the tension and the ones which are parallel to the applied pull. So shear can only take place on the planes which are inclined at

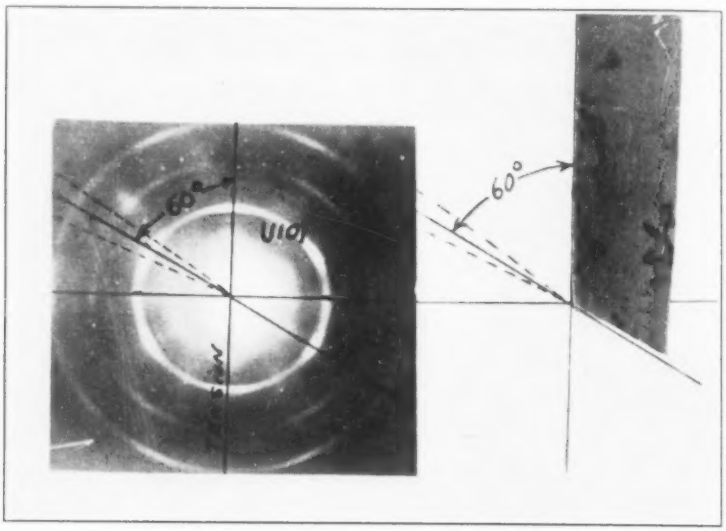

Fig. 22—X-Ray Diffraction Diagrams Prove that When a Test Piece is Ruptured in a Tensile Machine Failure Takes Place by Shear on the (1 1 0) Planes Making an Angle of Approximately 60 Degrees With the Axis of a Test Specimen. The Test Specimen After Fracture is Shown to the Right of the X-Ray Diagram and the Angle is 60 Degrees Proving the Correctness of the X-Ray Analysis.

60 degrees to the tension axis. There are two sets of these planes inclined at 60 degrees to the axis of pull, but failure oftentimes occurs only on one set, (see Fig. 22).

In some strip steel specimens tested in tension the slip lines develop all over the surface at 60 degrees to the axis of pull. Shortly before failure, one can actually see the shear continue on one group of (1 1 0) glide surfaces, then suddenly stop, and begin glide on another set, making an angle of 120 degrees with the old set, and continue to slip on this set of shear (1 1 0) planes until failure.

Sometimes shear failure takes place on both sets of (1 1 0) shear planes simultaneously. The test strip then has a fish tail fracture.

It should be pointed out that the shape of the strip steel specimen influences the type of fracture, and in order to obtain the kind described, a flat strip must be used, having a rectangular cross section.

On examining the X-ray diagram (Fig. 22) made at the fracture, one can see that the spots are spread over a considerable angle (about 10 degrees of arc). The (1 1 0) shear planes do not make an exact angle of 60 degrees with the axis of tension but vary about 5 degrees from the ideal position. This accounts for the variation in the angle of fracture with the axis of tension. However, it is certainly remarkable how close to 60 degrees the angle is when the angle is measured on a large number of strip steel specimens tested in this way.

The relationship between the shear fracture on the test specimen, and the X-ray fiber diagram showing the orientation is clearly shown in Fig. 22.

Compression tests were also made on a low carbon steel, 0.375 inch diameter and 0.375 inch long. After upsetting the specimen it was 0.850 inch thick, but it did not fracture. The X-ray structure was interesting because the fiber diagram was similar to that developed in wires pulled to failure in tension.

New Orientations Developed During Heat Treatment and Their Stability on Further Cold Rolling

Several years ago the writer (20) found that when a low carbon strip steel is given an area reduction of 96 per cent and heat treated, a new recrystallization orientation is developed. In the cold-worked strip the [1 1 0] coincides with the rolling direction, with the (1 0 0) and (2 1 1) planes predominating in the rolling plane.

After heat treatment the (2 1 1) plane was found to coincide with the direction of rolling instead of the [1 1 0] direction; heat treating favoring the formation of annealing twins. This interpretation is in line with the results published by L. W. McKeehan and others. For example, Mathewson and Edmunds obtained Laue patterns of a silicon steel which contained Newmann bands, and from an examination of the diffraction spots, concluded that twinning had occurred on the (2 1 1) planes. The twin bands can only be produced by shock. Slow straining would result in slipping on the (1 1 0) planes.

It was of interest to determine by experiment whether or not this new recrystallization fiber was stable when further cold-rolled. Ac-

cordingly, a strip steel known to have this recrystallization structure was given an area reduction of 50 per cent. The X-ray proved that the new orientation was retained but some of the grain fragments also were reoriented in some other direction (see Fig. 24). This means that the twins are metastable.

A similar structure is developed in copper but the new orientation is not changed when given an area reduction of 75 per cent. (Compare Figs. 24 and 25.) Apparently the recrystallization fiber of the copper is more stable or else the transformation into the new

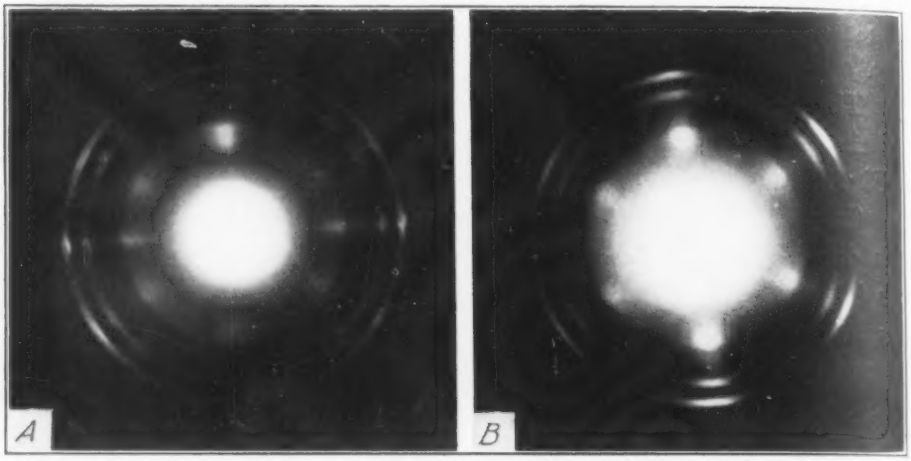

Fig. 23A—X-Ray Diffraction Diagram of a Low Carbon Strip Steel Given an Area Reduction of 99 Per Cent, X-Rayed in the N Direction.

Fig. 23B—X-Ray Diffraction Diagram of a Low Carbon Strip Steel, X-Rayed in the T Direction. Experience Shows that the Preferred Orientation Develops More Readily in the T Direction than in the N Direction.

orientation structure is more complete in the copper strips than in the steel strips. This is a very interesting phenomenon and one that should receive further consideration.

Another interesting recrystallization re-orientation was recently found in a 0.55 per cent carbon steel. The fiber structure, as yet, has not been determined, however, it must be due to some kind of twinning. A hot-rolled strip was cold-rolled from 0.062 to 0.031 inch and X-rayed in the (P), (N) and (T) directions. (See Fig. 26.) No preferred orientation could be found in the (P) and (N) directions, but a strong fiber structure appeared in the (T) direction (see Fig. 26a). The object of the experiment was to determine the temperature required to remove the directional properties in the (T) direction. When this strip was heat treated at 1300 degrees Fahr. for thirty minutes and air-cooled, a new preferred orientation

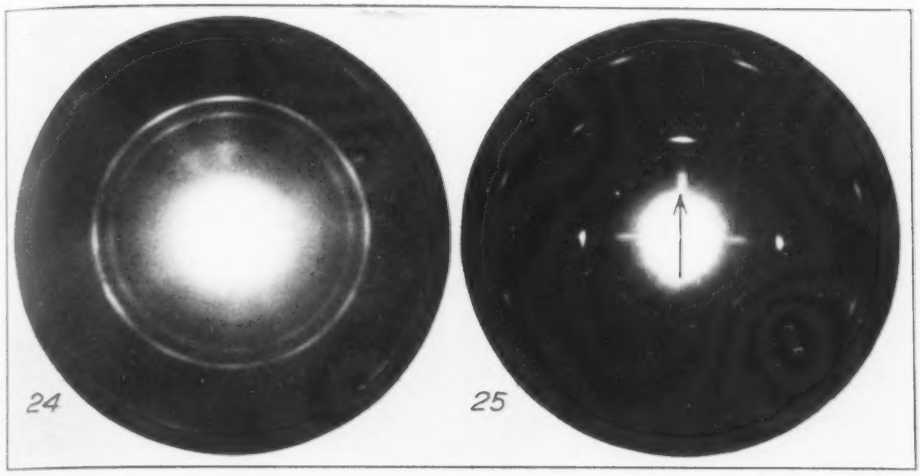

Fig. 24—X-Ray Diffraction Diagram of New Type of Fiber Structure Found in Low Carbon Strip Steels. Compare with Fig. 11 Which is Also Made of Low Carbon Strip Steel.

Fig. 25—X-Ray Diffraction Diagram of a Copper Strip Having Recrystallization Structure Cold-Rolled 75 Per Cent and Showing Great Stability.

appeared (Fig. 26b) (X-rayed in (T) direction). When this same cold-rolled strip was heat treated at 1400 degrees Fahr. a random grain structure resulted, Fig. 26c. It is important to point out that if this strip steel had only been examined in the (N) direction, this new recrystallization would have escaped observation. It is also of importance to note that this new recrystallization orientation appeared in a high carbon strip steel given an area reduction of 50 per cent. In the low carbon strip steel, the fiber structure was found in the (N) direction, but in the high carbon steel in the (T) direction.

Time Required to Remove Directional Properties in High Carbon Wires and to Complete Recrystallization of Cold-Worked Strip Steels

In many technical applications, it is important to know whether or not the heat treatment has removed the directional properties of the grains. When the area reduction does not exceed 50 per cent, proper heat treatment will usually remove directional properties. However, this is not always true.

The next question is how long does it take to remove directional properties? For the purpose of illustration a 0.60 per cent carbon steel wire was given area reductions of 35, 50, and 75 per cent through wire drawing dies. The wires were all ⅛ inch thick when

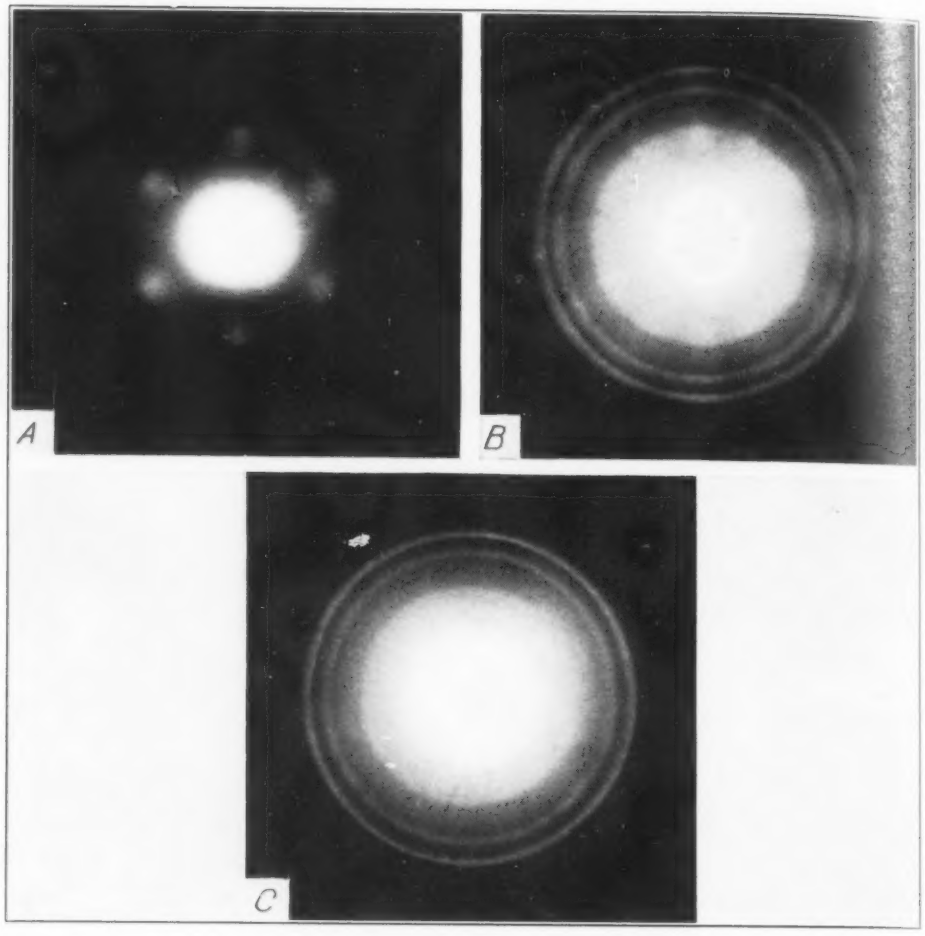

Fig. 26—X-Ray Diffraction Diagrams all Made in the T Direction. A—As Cold-Rolled. B—Heat Treated Below the A_3 Point. C—Heat Treated Above the A_3 Point. Note that a New Recrystallization Structure Fiber Has Been Introduced When Annealed Below A_3.

finished. They were heat treated at 1600 degrees Fahr. which is about the A_3; and held at this temperature for 30, 45, 60 and 150 seconds.

Each specimen was fastened to the end of a thermocouple and placed in the furnace and held at the temperature for the time specified. The results are summarized in the following table:

	Directional Properties After Heat Treatment		
1600 Degrees Fahr. Time at Temperature	Area Reduction 35 Per Cent	Area Reduction 50 Per Cent	Area Reduction 75 Per Cent
30 seconds	Very Weak	Weak	Strong
45 seconds	Very Weak	Weak	Strong
60 seconds	Very Weak	Weak	Strong
150 seconds	Very Weak	Weak	Strong

It was interesting to find residual orientation retained in all these wires after heat treatment. The intensity of the residual fiber struc-

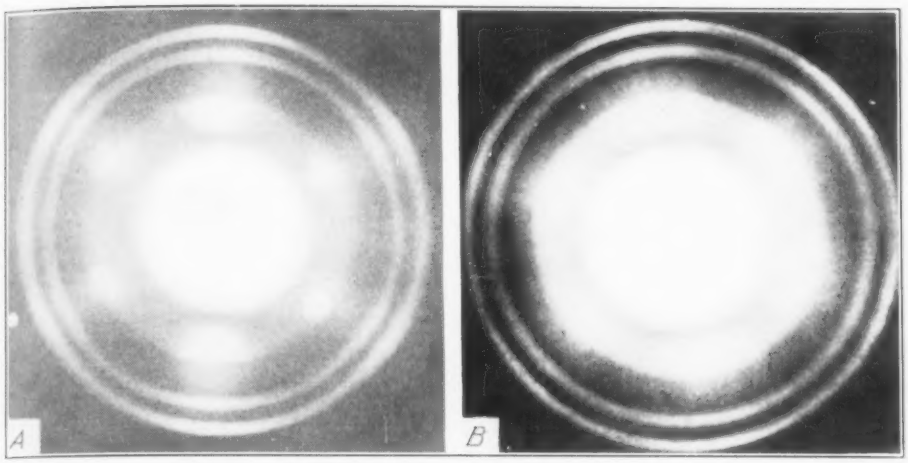

Fig. 27A and B—Proves that it is Sometimes Very Difficult to Remove the Directional Properties Even When Heated Above the A_3 Point.

ture increased with the area reduction. The time at temperature did not seem to have an effect on the amount of residual fiber retained. (See Fig. 27.)

One would think that the directional properties would be removed since the temperature was above A_3. The space lattice of the grains change from the face-centered cubic into the body-centered cubic space lattice when passing through A_3 on cooling, and since this transformation occurs twice, the directional properties should be entirely removed, regardless of the area reduction.

Auto valve springs made from such wires oftentimes fail in service and without exception residual directional properties are found. Auto valve springs which did not fail in service, and examined with X-ray did not show any residual fiber, or if present, it was very weak. When the grains are strongly oriented they have a preferred direction of weakness, while in the structures having the grains oriented at random they do not have this inherent weakness. The fracture usually makes an angle of 45 degrees to the wire axis, proving that it is a brittle fracture on the (1 0 0) cleavage planes. The result of the X-ray analysis clearly shows that residual fiber retained after heat treatment influences the fatigue properties of springs. It should be pointed out that in some structures it might be desirable to retain these directional properties because of the beneficial effect on certain physical properties.

Although the X-ray can also be used for testing the directional properties in flat rolled strip, it seems that this is a rather laborious

task and can be determined much more rapidly by a cup test. The X-ray should be used only to determine the orientation of the grains, and the cup test in production testing.

Time Required for Complete Recrystallization in Strip Metals

The X-ray can be used to advantage in studying the time required for complete recrystallization of cold-worked metals, and also the type of structure developed if the time is varied. Some alloys recrystallize rapidly, and others very slowly. The rate at which metals recrystallize depends on a number of factors such as:

1. Chemical composition.
2. Amount of hot and cold working.
3. Temperature of heat treatment.
4. Rate of heating and time at temperature.
5. Gage of the strip.

The amount of cold work a metal has received influences the temperature at which recrystallization starts. For example, a strip steel given an area reduction of 10 per cent fails (Fig. 28a) to recrystallize when heat treated at 1100 degrees Fahr. for one-half hour. When given an area reduction of 20 per cent it is partially (Fig. 28b) recrystallized. As the amount of plastic deformation is increased, the time at temperature is greatly decreased i.e., a strip of low carbon strip steel given an area reduction of 80 per cent is completely recrystallized in one minute at a temperature of only 1200 degrees Fahr.

A strip of brass which was cold-rolled severely, was heat treated at 1600 degrees Fahr. for various times. One specimen was placed in the furnace and removed as soon as the specimen reached the temperature (Fig. 28c) while another was held at this temperature for about 30 seconds (Fig. 28d). These X-ray diagrams prove that grain growth has occurred in the specimen with increased annealing time.

A strip steel which was given an area reduction of 98 per cent (0.09 per cent carbon) and heat treated at 1000 degrees Fahr. did not completely recrystallize in 15 minutes. The recrystallization structure (Fig. 24) is the twin structure already described. Another strip steel cold-worked 80 per cent completely recrystallized in one minute

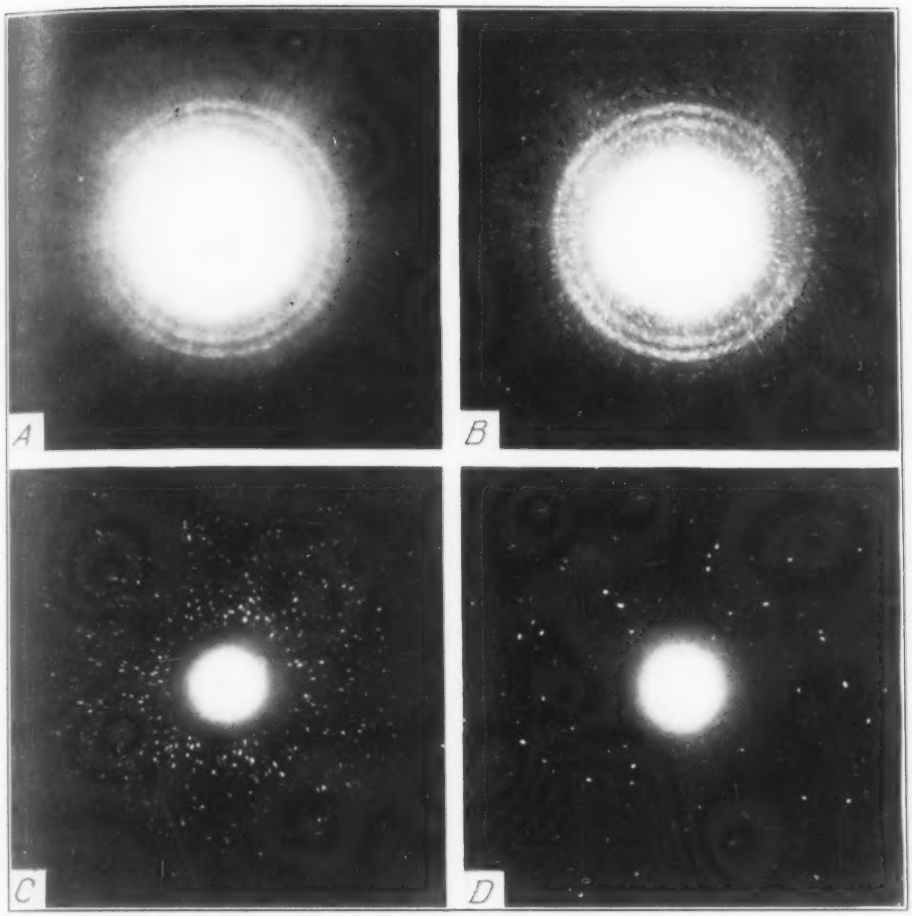

Fig. 28—X-Ray Diffraction Diagrams of Cold-Worked Metals Having Received Various Heat Treatments. A—Heat Treated at 1100 Degrees Fahr. for ½ Hour, No Recrystallization. B—Low Carbon Strip Steel, Cold-Rolled 20 Per Cent and is only Partially Recrystallized When Heat Treated at 1200 Degrees Fahr. for 1 Hour. C and D—Shows that the X-Ray can Follow the Progress of Grain Growth in Metals.

at 1000 degrees Fahr. One would expect that the strip which was cold-rolled the greater amount would respond more rapidly to recrystallization, the X-ray proves that this is not necessarily true, but depends on what is taking place during the process of recrystallization.

From an economic point of view, the accurate determination of the time required to completely recrystallize a cold-rolled or hot-rolled metal is very important. Any saving in time in the heat treating operation means increased production at a lower cost and the X-ray method offers a distinct advantage in studies of this kind, especially in complex alloys.

Directional Properties Imparted to Metals Given Small and Large Plastic Deformations

Cold rolling a strip steel sufficiently introduces directional properties. The X-ray diagrams shown in Fig. 21 do not indicate any evidence of directional properties when this strip is given an area reduction of 15 per cent. This amount of cold rolling is large enough to change the physical properties with a cross grain. These X-ray diagrams would imply that the physical properties with and cross grain are the same, but actually they differ. The directional properties are not found in the direction (N). In a recent paper (22) it was clearly demonstrated that X-ray diagrams made by transmitting the X-ray beam normal to the surface of the strip will not detect the existence of the preferred orientation of the grains in sheets given small reductions. The directional properties can only be found when X-rayed in the (T) direction. When the grain size is small enough (0.004-inch diameter or under) and area reduction of 10 per cent can introduce preferred orientation of the grains.

Strip steel must be drastically deformed before any orientation can be observed on the X-ray diagrams made by transmitting the X-ray beam normal to the surface. In Fig. 23a for example, is an X-ray diagram of a strip steel reduced 99 per cent by cold rolling. This single X-ray pattern is sufficient to prove that the [1 1 0] direction of the grain fragments lie in the direction of rolling. However, it is not sufficient for determining which planes lie in the surface of the strip. Several have been definitely (23) reported to lie in the rolling plane, namely the (1 1 0), (2 1 1) and (1 1 1) planes.

When low carbon strip steel specimens given an area reduction of 70 per cent are examined in the transverse (T) and parallel (P) one finds that both (2 1 1) and the (1 0 0) lie in the plane or rolling. The X-ray diagram (Fig. 23b) shows definitely that the positions of the (1 0 0) maxima require a (2 1 1) as well as (1 0 0) planes to lie in the rolling plane.

The difference in the fiber structure of cold-rolled strip in the (P) and (T) directions is sufficient to explain the difference in the physical properties with and cross grain of cold-rolled strips or sheets since the (P) and (T) fiber diagrams differ. If the fiber diagrams were the same in the (P) and (T) directions, then certain physical properties would be about equal. This can be actually accomplished by the method of cross rolling, i.e., alternating the passes by turning

the sheet 90 degrees after each pass. The older theory of orientation was based on all symmetrical orientation of the grain fragments, and this did not explain the variation in the with and cross grain properties of the strip. The physical properties with and cross grain are not in a definite ratio, but vary, and this can also be explained by the fact that the number of (1 0 0) (2 1 1) in the surface of the sheet along with the less prominent planes (1 1 1), (1 2 3), (3 1 1), etc. vary from strip to strip. A number of factors influence this, a few of them being, 1. chemical composition, 2. grain orientation in hot-rolled strip.

In a cold-rolled low carbon steel alloy examined by the reflection method already described, the (1 1 0) and (2 1 1) planes were all strong in the rolling surface, but the [1 1 0] direction was found to coincide with the rolling direction in all the specimens.

The Effect of Bending and Twisting on Annealed Wires

Torsion Test—Very little information is found in the literature concerning the crystal structure of steel wires and other metals which have been twisted or tested in torsion.

Ono (24) examined the torsion structure of copper, aluminum and iron wires, and found that these structures have the grain fragments oriented in about the same way as a wire which has been drafted through wire drawing dies.

In iron for example Ono finds that the (2 1 1) and the (1 1 0) planes are both parallel to the wire axis after twisting. His conclusions would infer that twisting has the same orienting effect as drawing through a wire drawing die.

In the first place the fiber structure developed at the point of fracture, is not nearly as intensely developed as in a wire given an area reduction of about 40 per cent.

A torsion test is usually made on a gage length of about 8 or more inches. For a time the twisting is uniformly distributed over the entire test piece, but with continued twisting the twists suddenly localize in one spot, and failure by shearing occurs rapidly, with only a few additional twists. The twisting action which was distributed uniformly over the entire section, is suddenly localized in one section, and the load is concentrated in this section, and failure soon takes place. X-ray fiber diagrams made in various directions near the point of fracture clearly show that the orientation of the grain frag-

ments is quite complex, although it is interesting to find many grains oriented with the [1 1 0] direction along the axis of the wire.

The following X-ray investigation was made of annealed wires having received various amounts of twisting, until failure, with the purpose of determining the intensity of the fiber structure developed at the point of failure.

A small coil of low carbon rod was drafted from 0.375 to 0.290 inch and heat treated at 1200 to 1250 degrees Fahr. for three hours, and slowly cooled in the furnace. The annealed wire was cut into 10-inch lengths, for the torsion tests. X-ray examination of a few specimens of the annealed wire proved that the wire was completely recrystallized, and the grains were oriented at random, and the perfection of the grains varied.

The wires were mounted in a torsion testing machine, with the jaws 8 inches apart. The first wire was given one twist, the second three twists and the third five twists, and so on until failure. For this gage and thickness of wire, 38 twists were required before failure.

All of the specimens for X-ray examination were removed from the center of the specimen and parallel to the wire axis.

Only one twist was sufficient to fragment the grains considerably (but not sufficient to introduce directional properties). This could be easily verified by comparing the X-ray diagrams of the annealed and twisted wires. The X-ray diffraction diagram of the wire given three twists showed further evidence of grain fragmentation, but without any evidence of preferred orientation.

Five twists further fragments the grains and preferred orientation begins to appear. With continued twisting the fragmentation continues and orientation increases up to the point of failure.

The X-ray specimen of the wire which failed in torsion was made at the point of fracture. The fiber pattern was not as sharply developed as in wires having received an area of 40 per cent. This is a very interesting observation, because one might expect the preferred orientation at the fracture to be more intensely developed.

Failure therefore, is attributed to the sudden localization of the shear stresses in one point in the wire, and slip finally takes place on a limited number of shear planes of the (1 1 0) type when it is a ductile fracture and on the (1 0 0) planes when it is a brittle fracture.

Fiber Structure Developed During Bend Twisting

It is of value to know whether bending a wire successively

through alternate right angle bends develops the same type of fiber structure as a wire tested in torsion or by stretching.

The bending test is often times used as a test for ductility in strip and wires, and it is important to know the changes which occur in the grain structure as the metal is bent.

The wire used in the torsion experiment was also used in this experiment, so the results are comparative.

A wire was placed in a vise having special jaws suited for tests of this kind. It is considered to have received one bend when it is bent through an angle of 90 degrees and returned to its original starting position. The next bend was made in the opposite direction, and this process was repeated until the wire failed. The first wire was given one bend, the second two bends and so on. A specimen for X-ray examination was prepared from each wire, parallel to the wire axis.

One bend is enough to introduce considerable grain fragmentation, but no orientation could be observed. The grains which were mosaic in the annealed wire are now subdivided into much smaller units, and the "block" fragments deviate more than before from each other.

Two bends further fragments the grains but their identity is not erased; three bends were required to start grain orientation; however, grain fragmentation is not complete. Five bends were sufficient to obliterate the identity of the individual diffraction spots, which means that the size of the "blocks" is becoming smaller and smaller with continued bending.

The wire failed at 10 bends. X-ray diagrams made of specimens removed near the fracture, however, did not exhibit intense fiber structure. The failure by bending is similar to torsion failures, from which one can conclude that the mechanism of the plastic flow in torsion and bending tests is similar.

Appendix

Hardening of Metals by Cold Working in Light of X-Ray Diffraction

When the yield point of a metal is exceeded, several important changes occur in the internal structure of the grains and the physical properties. The tensile strength is increased and the ductility decreased. The electrical conductivity, resistance to corrosion, wear, density, and the magnetic properties are altered by cold working.

The rate at which a metal hardens depends upon the chemical composition and grain size, in general. Cold working modifies some physical properties more than others, while some remain practically unchanged. The modulus of elasticity of steel is 30,000,000, cold working does not change it, the space lattice also does not change. On the other hand the yield point is markedly altered, and the density changed only slightly.

Regardless of the magnitude of the changes in the physical properties by cold working all are related to some modification which has occurred within the grains.

Here, however, only the changes which contribute to increase the hardness of a metal when plastically deformed will be considered.

Some metals can be plastically deformed more than others without becoming very brittle. Copper, when drastically cold-worked, retains a surprising amount of ductility, while high carbon strip steel deformed the same amount appears brittle. A brittle metal obviously cannot be deformed plastically. A metal capable of being deformed plastically may be likened to a viscous fluid, becoming more and more viscous as cold working is continued. The various items outlined on page 981 will now be considered in detail.

Item 1—Ewing and Rosenhain were the first to observe that when a polished metal specimen is slightly deformed, huge blocks are displaced relative to each other (see Fig. 6A). Slip was not observed to occur on every possible glide plane, from which it would appear that the slip planes (potential) do not have the same shear strength. See page 980 of the text. Any theory which attempts to explain the mechanism of plastic flow must include this very important discovery by Ewing and Rosenhain.

In Fig. 6B an X-ray pattern of a single grain given a slight plastic deformation is shown. The single spots of the undeformed grain shown in Fig. 4A are replaced by split diffraction spots. This means that large lattice fragments have been displaced by this small deformation. The thing that remains to be established is whether the displacement is one of pure translation, or whether rotation also accompanies it, when the deformation is less than 0.5 per cent. When the metal is deformed 2 per cent, considerable rotation of the lattice "blocks" takes place. Strip steels containing several per cent silicon are well suited for an investigation of this kind, since the "blocks" are usually perfectly aligned within the grain, and in addition the "blocks" are quite large.

Item 2—Continued cold working fragments the grains into smaller units, and these are definitely oriented relative to the direction of working. Only a small (19a) area reduction is required to orient the grains. The greater the cold working the more perfectly developed a fiber structure, but the severest reduction fails to attain the ideal orientation.

In Fig. 11 the X-ray diagram of a low carbon strip steel given an area reduction of 97.5 per cent is shown; as can be seen, this drastic deformation fails to orient the grain fragments in the ideal position.

In an annealed metal in which the grains are initially oriented at random, cold working fragments the grains into smaller units (blocks). The mechanism of deformation is one of rotation and translation on the glide surfaces. The displaced blocks in addition are oriented definitely in the direction of working.

Tammann, some time ago, developed a strain hardening theory, in which gliding was considered to be entirely a translatory motion. This would mean that the "blocks" after deformation retained the same orientation as the original grain. Bain and Jeffries[25] proved this to be untenable. They proved that plastic deformation of single ductile crystals developed a mixed orientation. This can only be explained if we assume that the block movements are not only translatory but rotational as well.

Item 3—It is well known that if the cold working of a metal is continued, increased resistance to further deformation occurs, plastic flow becoming more and more difficult. Although we know that metals deform by shear, no one can estimate what the resistance to glide should be at all stages of deformation. Our knowledge of the cohesive forces in metals is very incomplete. Duwez,[26] believes that gliding in crystals takes place between the "blocks" of the secondary structure, and discusses the theoretical aspects of this important problem.

M. Kornfeld[27] from theoretical considerations shows that a large portion of the latent energy induced by cold working is located at the interfaces of the dislocated blocks; only a small portion of the

[25]E. C. Bain and Zay Jeffries, "Mixed Orientations Developed in Crystals of Ductile Metals by Plastic Deformation," *Chemical and Metallurgical Engineering*, Vol. 25, p. 775, Oct. 26, 1921.

[26]Pol Duwez, "On the Plasticity of Crystals," *Physical Review*, Vol. 47, March 1935, p. 494.

[27]M. Kornfeld, "The Structure of a Deformed Crystal and Recovery Phenomena," *Phys. Zeit du Sowjitunion*, Vol. 7, No. 5/6, 1935, p. 608-18.

energy absorbed causes lattice distortion. Joffé believes that the strengthening or resistance to gliding on the slip planes is due to an increase or change in potential energy. Jeffries and Archer were the first to put forward a satisfactory theory, which was consistent with the facts, namely their well known slip plane-interference theory.[28] W. Fuller has also shown that strain hardening does not disturb the space lattice appreciably. All of these theories and many more have contributed to a better understanding of shear hardening of metals.

Item 4—On pages 977 to 979 of the paper, a somewhat different interpretation of radial asterism is given and which is in better agreement with the observed experimental facts. This interpretation is also in excellent agreement with the results of Bain and Jeffries[25] obtained in 1921 on deformed single crystals. The asterism proves that rotation of the "blocks" has taken place during plastic deformation, which means that the displaced "blocks" assume an orientation differing from each other as well as from the original grain.

To illustrate this consider the X-ray diffraction diagram of the normalized strip shown in Fig. 21A. The grains in the heat treated strips are free of asterism. After this strip was given an area reduction of only 5 per cent, asterism appears on the X-ray pattern of the deformed metal. This means that the "blocks" within the grains have been displaced. The asterism now present can only be accounted for if we consider that during the process of fragmentation, the "blocks" were rotated on the glide planes. The deviation from block to block is gradual as the uniform texture of the asterism suggests. In some instances the asterism consists of striations parallel to each other, in such deformation structures the blocks are quite large, and deviate considerably from each other (adjacent blocks do). When the radial streaks (asterism) are of uniform density then the blocks are either very small, or deviate gradually from each other, or it may be due to a combination of both. However, the total block displacement is quite large. (Several degrees of arc).

It can also be appreciated that if the gliding during deformation is only translatory no asterism would be possible, since the displaced atom layers would have exactly the same orientations after deformation as the original grain.

[28]The Science of Metals—Jeffries and Archer—p. 405, McGraw-Hill.

[29]W. Fuller, "Strain Hardening of Metals by Cold Work," Die Umschau in Wissenschaft und Technik, Vol. 38, July 1, 1934, p. 528-30.

The "asterism" found on all X-ray diagrams of cold-worked or heat treated metals is therefore conclusive proof that rotations of the blocks has accompanied the deformation.

As a matter of fact the X-ray diagram never did show directly that the grain fragments were translated during deformation. It can only show (1) That the grains were fragmented into smaller units.

(a) Block displacement

(b) The "blocks" internal structure consisting of particles ranging from colloidal up.

(2) That the "blocks" rotated on the glide surfaces and cohesion is maintained. (3) That the grain fragments (blocks) are oriented in a definite direction.

Item 5.—Pulsifer has shown that in a wire given an area reduction of 99.95 per cent by drawing through wire drawing dies in 31 passes (Fig. 29), the identity of the blocks is retained even after such

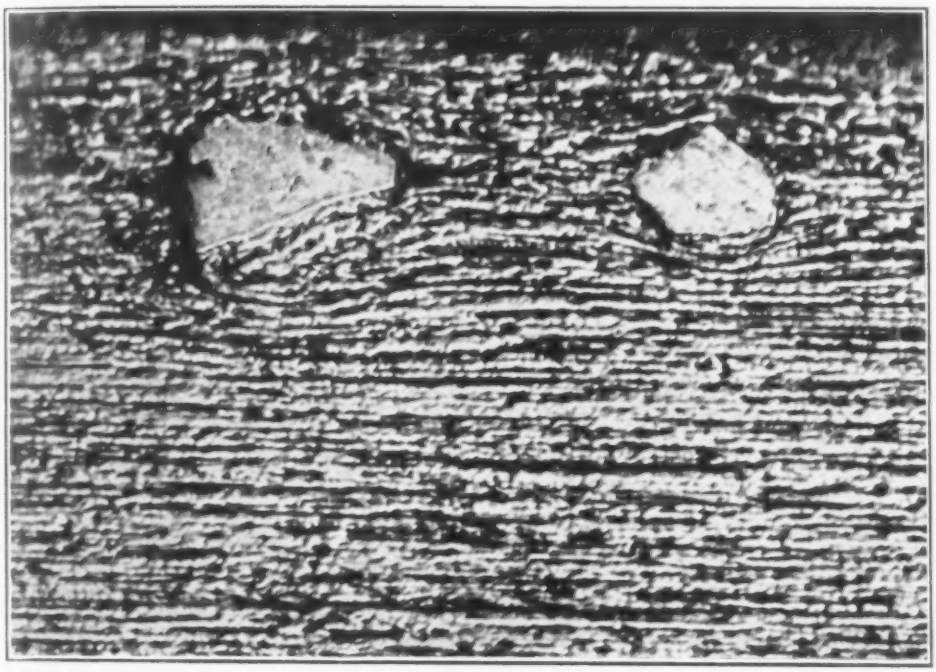

Fig. 29—Photomicrograph of Wire 0.0049 Inch in Diameter Given an Area Reduction of 99.95 Per Cent by Drawing Through Wire Drawing Dies in 31 Passes (Pulsifer)

a drastic plastic deformation. The blocks are only 16×10^{-5} m. m. and under in magnitude. Objections might be raised against the possibility of being able to measure the true size of the blocks from the "fine structure" diagrams, since various reagents give different fine

structure patterns. However this is not so important at the present time; what is important is that Pulsifer was able to prove that the "blocks" retain their identity even after the most drastic deformations.

The X-ray also shows that the "blocks" retain their identity after severe deformations, for the diffraction lines do not broaden as much as would be expected. The structure of the X-ray diffraction lines does not change much after the deformation exceeds 90 per cent. In fact the $K\alpha$ doublet of molybdenum radiation reflected from the surface of a severely worked strip steel is resolved, but the $K\alpha$ doublet lines are broadened slightly, however, the intensity of the higher order lines is usually greatly reduced.

Items 6, 7 and 8—To explain the increase in hardness and strength and loss of ductility of metals when cold-worked, G. T. Beilby in 1911 conceived the idea that plastic deformation caused amorphous material to form on the slip planes and that the grains were held together by an amorphous cement. This established a good working hypothesis from which all of the present day theories have been derived.

He also postulated that drastic cold working completely reduced the crystalline grains into an amorphous state. However, after X-ray diffraction studies were made of drastically deformed metals, at a later date, Beilby's hypothesis was found to be untenable. X-ray diffraction has conclusively shown that deformed metals retain their crystalline structure. Some still believe that a small amount of the deformed crystalline debris is reduced to the amorphous state, but the X-ray cannot detect it.

However Jeffries and Archer have shown that the hardening of metals by cold working does not require an amorphous phase to explain the facts.

What holds the grains together at the grain boundary is still a matter of great speculation, however a great many authorities believe that the cohesion at the grain boundaries is accounted for if one assumes the presence of an extremely thin film of amorphous cement. However it would seem that the facts can be explained by considering cohesion due entirely to molecular forces, since the surfaces of the adjacent grains approach each other very closely (order of magnitude 2-5 10^{-8} m.m.).

Electron diffraction methods are better suited to investigations of this kind, since the electron penetration can be confined to only a

depth of a few atom layers, thus the surface structure of crystals can be studied.

French[30] studied the polished surfaces of copper, stainless steel, chromium, etc., with electron diffraction apparatus. He found the Debye rings to broaden considerably. Randall and Rooksby thought that his results could be best explained, if the particle size of the displaced lattice fragments is considered to be very small in order to account for the line broadening.

In polishing operations, plastic flow only penetrates the metal a very small distance (several hundred atom layers at most). The surface cells are therefore severely worked and these "blocks" are displaced from the original lattice by the process of gliding (translational and rotational) but molecular cohesion is maintained, in fact it is increased!!

X-Ray Structure of Amorphous and Crystalline Matter

In amorphous matter the atoms are arranged haphazardly, for this reason such substances do not possess any mechanical weakness in some particular direction, since the atoms are not arranged on a space lattice. Such substances are not only brittle but are highly viscous, and cannot be deformed plastically. In addition these vitreous substances are very hard. Glasses are vitreous substances devoid of any crystalline properties. Amorphous materials when X-rayed produce X-ray diagrams characteristic of such substances, and differ from those obtained from crystalline matter.

X-ray diffraction diagrams of polycrystalline materials are always characterized by many sharp diffraction lines, and the same line may appear in three or more orders. Amorphous materials give X-ray diagrams having only one or two broad bands, in every respect similar to a liquid. In fact a vitreous substance may be considered similar to an undercooled liquid.

Randall, Rooksby and Cooper[31] examined a number of substances with the X-ray, which could be obtained both in the crystalline and vitreous form. The crystalline modification as expected gave X-ray diagrams having numerous sharp rings, while the corresponding glasses had one or two broad bands. In their opinion the corresponding patterns were similar, in that the broad band of

[30]R. C. French, *Proceedings*, Royal Society, Vol. 140, 1933, p. 637.
[31]*Nature*, Vol. 125, 1930, p. 458.

the vitreous substance appeared in approximately the same position on the X-ray diagram as the sharp ring of the crystalline modification. Microphotometer records indicated the broad bands to consist of several lines, indicating at least that some intimate relationship exists between the vitreous and crystalline states of matter.

Crystal Size and Line Width

The colloidal chemists have made some valuable contributions in X-ray diffraction investigations in an attempt to correlate line broadening to particle size.

Scherrer[32] was perhaps the first to show that as the particle size of the colloid became smaller and smaller the diffraction lines became broader and broader.

Particles only 1μ (10^{-3} m. m.) will give a sharp line diffraction pattern. When the particles are smaller than 10^{-5} m. m. the Debye rings broaden.

Scherrer studied the particle size of colloidal gold, and found that the line broadening increased as the colloidal particle became smaller and smaller. He found that the average particle size could be calculated by the use of the following equation devised by Debye and Scherrer.

$$B = 2\sqrt{\frac{\text{Log}e^2}{\pi}}\ \frac{\lambda}{D}\ \frac{1}{\cos\frac{\Theta}{2}} + b$$

B = angular width of halo, measured between points at which the intensity has been reduced to its half max value.
Θ = Bragg Angle.
λ = Wave length of X-ray employed.
b = minimum breadth of halo determined from dimensions of X-ray beam and the size of powder capsule.

Scherrer found the smallest particle of the colloidal particle to be about 18.6 A. U. This particle was only about 4 unit cells in diameter (Au — A_{100} = 4.07 A. U.)

Randall[33] (page 30) studied a series of MgO powders having various particle sizes. Sharp lines were obtained from particles in

[32]Nach Göttingen Gasellschaft, 1918, p. 98.

[33]Randall, "The Diffraction of X-rays and Electrons by Amorphous Solids, Liquids and Gases." John Wiley & Sons.

10^{-3} m.m. but as the size decreased the lines broadened. He also points out (page 48) that the lattice spacing of a crystal or particle depends upon the size. (Lines do not appear in the same average position but are shifted.)

Others have considered line broadening to be associated with lattice distortion (due to elastic stresses) but the colloidal chemist has definitely shown that it is due to particle size. If line broadening is due to both then it will become a difficult task to separate one from the other.

Prof. E. N. da C. Andrade, (7a) states that particles containing less than 8 atoms cannot maintain themselves in a cubic space lattice and that 100 or even 1000 times that number will have a tendency to change not only the dimensions but the symmetry of the space lattice as well. It is now believed, and there is good evidence to support it that when the number of atom planes are not large in number the tendency is to increase the planar spacing. The results of X-ray analysis which have been derived from the investigation of amorphous and colloidal substances, can be directly applied to the study of cold-rolled metals.

The structure of the lines of the (reflection) X-ray diagrams of cold-worked metals tells us something about the internal structure of the displaced blocks. From the nature of the diffraction lines found on X-ray diagrams of cold-rolled strip steels given drastic reductions, one finds that the lines are not broadened to the extent that the Kα doublets are not obliterated, in fact they are found to be resolved even after the severest reduction. The doublet lines are in themselves quite diffuse and the intensity reduced. The texture of the lines can be accounted for if we consider that a portion of the particles of which the blocks are built is colloidal but the greater part are much larger (in the range 10^{-5} to 10^{-4} m.m.)

The decrease in the intensity of the lines can be in part accounted for if we consider the absorbed energy due to the cold working to have profoundly disturbed the electron distributions.

Item 9—When cold-worked structures are heat treated below recrystallization temperatures, the diffraction lines become sharper and the density increases. This sharpening of the diffraction lines can be accounted for if we assume that the particles within the blocks have coagulated, or realigned themselves. The process may also be one tending to smooth out the lattice parameter (to make them more uniform in size). The rather great variation in the particle size

within the block prior to tempering in view of Randall's work, means that the lattice dimensions vary considerably among the particles composing the block. After tempering the particle size tends to assume a more stable and uniform size, and as a consequence the glide surface is smoothened so to speak. The metal as a result of this becomes a little more ductile, and the hardness decreased.

The theory advanced here does not require the amorphous cement theory to explain why metals harden when cold-worked, but assumes the necessity of an inherent block structure to exist prior to deformation. Deformation displaces the blocks, and may reduce them to smaller units, but they never lose their identity. The blocks are small in comparison to the grain, but large when compared to the lattice dimensions. The blocks are the fine structure of the grains. In addition the "blocks" after deformation also have a fine structure, consisting of particles ranging from colloidal up to ones 10^{-5} m.m. or so in diameter. Tempering does not alter the position of the displaced blocks, since the asterism is not altered. Tempering however affects the internal structure of the blocks. The sharpening of the Kα doublets in severely cold-rolled strip steels indicates that the particles of colloidal size are increased in size, in fact the particles within the displaced blocks become more uniform in size.

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9. Zeit. Metallkunde 15 (1923) 60—Also see "Distortion of Metal Crystals," Elam Oxford, p. 169.
10. X-Ray Metallography in 1929, *Metals and Alloys,* (Reprint) p. 35.

10a. *Zeit für Physik,* Vol. 12, 1927, p. 58.

11. G. I. Finch and S. Fordham, *Physical Society Proceedings,* 48, p. 85-94, Jan. 1, 1936.
12. Note: D. M. McCutcheon, reports that the diffraction spots of the X-ray diagrams sharpen under stress. He applied an elastic stress of 10,000

pounds per square inch. These results certainly repudiate the ideal that asterism and diffuse diffraction spots in cold-worked metals means lattice distortion. *Metals and Alloys*, October, 1934, p. 230.
12a. Internal Stresses—Charles Barrett, *Metals and Alloys*, July, 1934, p. 156. (See Fig. 13, Laue photogram of an aluminum crystal before and after a stress relief anneal.)
13. Distortion of Metal Crystals, C. F. Elam, Oxford, p. 168.
14. *Philosophical Transactions*, 193, 1899, p. 353.
15. Crystal Structure of Magnesium Alloys—W. Schmidt Trans. by C. E. Doan, p. 149, *Metals and Alloys*, July, 1934.
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18. X-Rays Detect Twin Crystals in Alpha Iron, p. 1238-1245, TRANSACTIONS, American Society for Steel Treating, Vol. 18, July 1930-Dec. 1930.
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20. X-Rays Detect Twin Crystals in Alpha Iron, N. P. Goss, p. 1242, Fig. 5, American Society for Steel Treating, Vol. 18, July-December, 1930; TRANSACTIONS, American Society for Steel Treating, Sept. 1929, Effect of Annealing on the Grain Structure of Extremely Hard Rolled Steel.
21. Directional Properties in Rolled and Annealed Low Carbon Steel, A. Phillips and H. H. Dunkle, TRANSACTIONS, American Society for Metals, Oct. 1934 (reprint).
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DISCUSSION

Written Discussion: By Charles S. Barrett, Carnegie Institute of Technology, Pittsburgh.

On pages 1000 and 1001 of the paper, Mr. Goss concludes that strip steel tensile specimens break at an angle of 55 to 65 degrees from the tensile axis because of shear chiefly on (1 1 0) slip planes lying at this angle. This explanation for the angle of rupture and the conclusion that slip in iron is chiefly on (1 1 0) planes seems to be based on data that are entirely inadequate and that do not necessarily lead to the conclusions Mr. Goss has drawn.

Mr. Goss' reasoning seems to be something like this: since the diffraction pattern shows a concentration of (1 1 0) planes oriented so as to intersect the surface at 60 degrees, and since the fracture is at this angle, therefore the (1 1 0) slip planes at this angle must have been responsible for the fracture and must have determined its angle. But as Mr. Goss has mentioned earlier in the paper, and as Barrett, Ansel and Mehl have proved in another paper of this meeting, the numerous (1 1 2) and (1 2 3) planes of the crystals are also slip planes. Mr. Goss' failure to consider them and their angular positions invalidates his conclusions. He has likewise failed to take account of all planes whose normals do not lie parallel to the flat surface of the specimen, yet these should be at least as important as the ones he considers and will intersect the surface at a variety of angles quite different from the angles he discusses.

The problem of the angle of tensile fracture is a difficult one and has

already received considerable attention both from the experimental and theoretical point of view; this will be discussed by Mr. G. Ansel.

Recrystallization structures and the underlying reasons for them continue to mystify metallurgists, although some important advances in knowledge in the field have been made in recent years by Burgers, Dehlinger, Sachs and others. New theories are particularly welcome in this field for the right one seems always "just around the corner." I do not feel, however, that the theory offered in this paper will be of service. I refer to the theory on page 1002 that the recrystallization structure of rolled steel grows out of the deformation structure by the growth of annealing twins. If Mr. Goss will take a stereographic projection of a cubic crystal oriented in a manner appropriate to one of the grains in the rolled structure, and will plot the orientation of all the twins that can form from that grain, he will find that there are four new orientations thus produced and that none of these have the orientation he requires for his recrystallization structure. In other words, a twinning operation will not do what he assumes it will. As a matter of fact, both the rolling and the recrystallization structures in steel are so complex the question may well be raised as to whether a specification of them in such simple terms as Mr. Goss has used can really provide a rigorous test of any theory.

On page 1002 it is suggested that rolling a recrystallized sheet reorients some of the grain fragments because twins in the recrystallized sheet are metastable. This implies either that deformation twins form by rolling, which is not the case in low carbon steel, or that there is an untwinning process brought about by the rolling. Untwinning by mechanical deformation has been observed only in single crystals of tin under impact (the work of Chalmers) and would not be expected in the slow deformation of ferrite.

Written Discussion: By G. Ansel, Carnegie Institute of Technology, Pittsburgh.

The fracture of retangular strip specimens with no contraction in the width direction has been investigated by several men. The latest work is probably that of Bitter.[1] Working with rolled 4 per cent silicon sheet Dr. Bitter investigated the angle of fracture of strip specimens and its dependence upon the orientation of fiber structure. At liquid air temperatures he obtained (1 0 0) cleavage, as was expected, while at 200 degrees Cent. he obtained fracture at a characteristic angle. At room temperatures his fractures were a combination of (1 0 0) cleavage and characteristic angle fracture.

Dr. Bitter concluded that in the specimens exhibiting these so-called characteristic angle fractures the angle of break is more closely related to the direction of maximum elastic elongation rather than to the direction of maximum tension. He admits, however, that the conclusion rests on very scanty evidence.

Bauer, Göler and Sachs[2] reported data of the same nature on hard rolled sheets of copper and brass. The angles they obtained lie in the same range as those reported by Bitter, but their data has less scatter.

Simultaneous with the work of Bauer, Göler and Sachs an extensive in-

[1]F. Bitter, "On the Fracture of Fibered Iron Silicon Sheet," *Proceedings of the Royal Society*, A. Vol. 145, 1935.

[2]Bauer, Göler and Sachs, *Zeitschrift für Metallkunde*, Vol. 20, 1928, 202.

vestigation on the angle of fracture of rolled sheets of copper, aluminum, nickel, brass and electrolytic iron was carried out by Körber and Hoff.[3]

Their conclusions are of considerable interest. A ratio of thickness to width of 1 to 5 is necessary to get contraction while retaining the original specimen width, the contraction forming a definite angle with the specimen axis and the specimen finally breaking along this contraction.

Variation of the fracture angle with thickness of specimen and position of rolling direction was considerable. Specimens with the tension axis coinciding with the rolling direction showed in all metals the greatest fracture angle, the angle varying from 60 to 70 degrees. For other orientations the angle of tear oscillated irregularly about 55 degrees. With increasing ratio of width to thickness, the angle becomes generally smaller and approaches a mean of 55 degrees.

They concluded finally that the different lattices and the resulting different fiber structures did not seem to have an effect upon the fracture angle. They surmised that in these specimens rupture must occur by a macro mechanical process, and not, as Goss postulates, by slip on a specific atomic plane.

Körber and Siebel[4] have developed a theory which explains fairly satisfactorily the results obtained in all of the mentioned investigations. Up to the point of maximum load, the behavior of a round and rectangular specimen in the tensile test is the same. However as soon as contraction begins because of the fact that preceding deformation under tensile stress has exhausted some of the possibilities of flow of the material and also because the strip specimens are much wider than they are thick the directions of the principal stresses and the planes on which flow occurs are fixed. With the beginning of contraction in the direction of thickness, the change in form tries to occur only in those glide planes at 45 degrees to the tension axis and intersecting the rectangular cross section at 45 degrees to the principal stress directions lying in the specimen cross section. It is to be emphasized that these glide planes are not to be considered as crystallographic planes, but rather that they give the direction of displacement that leads to the observed change of form. Since the maximum load has been passed any further change in shape remains limited to these planes, which coincide with the first cross sectional weakening. These planes intersect the broadside of the rectangular specimen at an angle inclined at 55 degrees to the specimen axis. The contraction is finally ended by a shearing off of the specimen halves at this angle and the specimen fractures with a characteristic angle.

Körber and Siebel conclude that deviations of the fracture angle from 55 degrees are probably due to the influence of the rolling texture upon the angle. Their proof is for homogenous stress in an isotropic medium while the rolling texture is in some degree anisotropic. Deviations also might occur from the fact that the slipping mechanism in individual grains is of a crystallographic nature, slip in some grains occurring in directions deviating slightly from the specified macroscopic planes.

Written Discussion: By Howard A. Smith, Massachusetts Institute of Technology, Cambridge, Mass.

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[3]Körber and Hoff, *Mitt. K. Wilh. Inst. Eisenforsch.*, Vol. 110, 1928, p. 175.

[4]Körber and Siebel, *Mitt. K. Wilh. Inst. Eisenforsch.*, Vol. 110, 1928, p. 189.

This paper should be of considerable interest to physical metallurgists because it brings up again in connection with an important problem, the question of the relationship between internal strains and the interpretation of the X-ray diffraction photogram. In addition it also emphasizes the interpretation of the distortion of diffraction images in terms of the conception of block movement during plastic flow. This work of Mr. Goss is stimulating because it points the way to much new work and brings up the necessity for reinvestigating some previous work.

It seems to me that there is yet need for work from which it will be possible to correlate, quantitatively and with satisfactory care and completeness, the distortion effects in metallic crystals with the diffraction images. Such work should at first be limited to Laue photograms and perhaps single crystals. In planning such an investigation and in interpreting the results, one should have in mind as a probable interpretation, Mr. Goss' viewpoint of block and fibrous structure formation resulting from such an investigation which would directly serve two ends. First, it would form a satisfactory basis for the interpretation of photograms and, second, considerably more information would undoubtedly exist bearing on the nature and behavior of the block structure.

It is appreciated that considerable spread is represented in the physical and chemical characteristics and heat treatment of the alloys represented in this paper, and correctly so when ideas and methods are being blocked out. However, there are two types of alloys concerning which a detailed study would yield some very interesting information. I have in mind an investigation of the stability toward hot and cold work of an iron alloy containing about 6 per cent manganese, and a stainless alloy containing approximately 18 per cent chromium and 8 per cent nickel. The first of these transforms slowly during heat treatment, while the heat treatment transformation of the second is very sluggish. Both alloys transform, however, on working.

I have been considering possible explanations for the cause of the intense preferred development of the fiber pattern when the X-ray beam is transmitted in the (T) direction. I have not yet arrived at a satisfactory explanation and I am wondering what conclusions Mr. Goss may have regarding this.

Written Discussion: By R. E. Cramer, Materials Testing Laboratory, University of Illinois, Urbana, Illinois.

The writer's interest in Mr. Goss' paper is mostly in connection with the "fine structure" or internal structure of metallic grains. Mr. Goss is to be commended for bringing this subject up for discussion, as it is certainly an important phase of metallography which has received too little attention. Mr. Goss has also furnished a rather complete bibliography on fine structures but did not include one of the earliest publications on the subject, namely Bulletin No. 176 (1928), Engineering Experiment Station, University of Illinois—"A Metallographic Study of the Path of Fatigue Failure in Copper" by H. F. Moore and F. C. Howard. This bulletin showed the fine structure of copper up to 11,500 magnification. Those metallurgists who are more familiar with photomicrographs than X-ray patterns will regret that more micros of the fine structure of Mr. Goss' specimen are not available for comparison. This of course is asking too much of one author, as the material

presented represents a large amount of work. Mr. Goss has made a strong case for the "block structure" conception of metals in this paper.

Fine structure of metals has been a hobby with the writer since working with Mr. Goss several years ago on silicon steel. It would be well to point out that the original magnification of the two photomicrographs of Fig. 3 was 1500 reduced to 520 in reproduction. The accompanying Fig. A and B shows the fine structure in two micros of another specimen of silicon steel at 1000 magnification. Both were taken of the same area with the same etch, the only difference being in the focus. In Fig. A the fine structure boundaries appear light colored while Fig. B is a more natural focus. Some very interesting etching effects are often observed in the coarse grain boundaries of silicon steel after deep etching, especially a stair step effect shown to some extent at the top of these two micros.

Permalloy is another alloy which shows spectacular fine structures. It is an austenitic material and the fine structure seems to etch out more distinctly than in ferritic materials. Figs. C and D show two micros of permalloy at 2000 magnification deep etched in Merica's reagent.

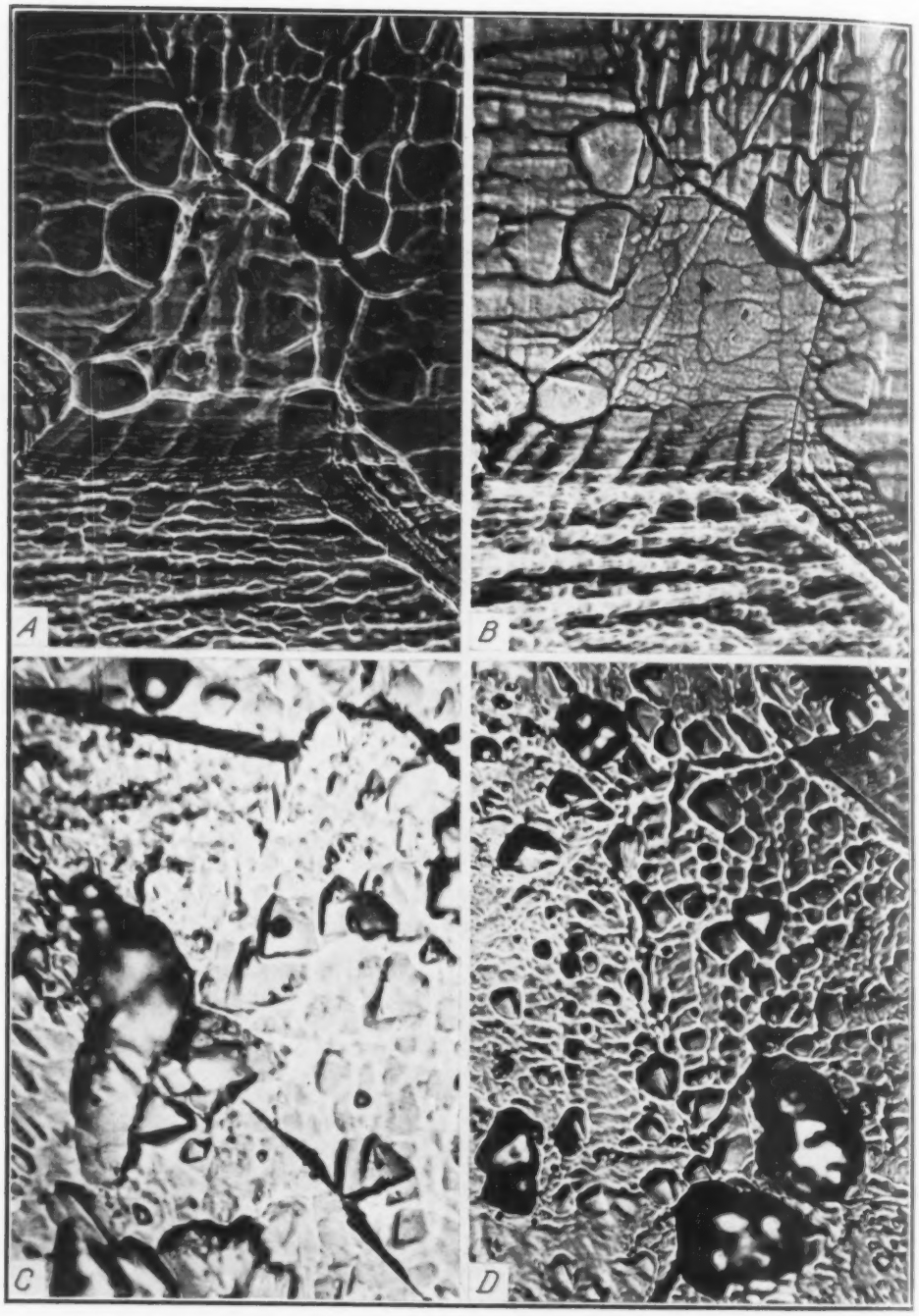
Figs. E and F show two views of the fine structure of Armco iron. The specimen shown in Fig. E is moderately deep etched and the fine structure is outlined but not completely revealed. The specimen shown in Fig. F is so deeply etched that only one grain could be brought into focus at a time. The blocks of fine structure seem to have curved surfaces.

It should be pointed out that fine structures are much more distinct under the microscope in metals which have been severely cold-worked followed by recrystallization than in hot-rolled materials. It is the writer's opinion that the boundaries of the fine structure represent scars along the slip planes caused by cold working of the metal which have not been completely eliminated in the recrystallization process. The writer has seen traces of such structure in deeply etched specimens of thin hot-rolled strip, but they were never distinct enough to photograph satisfactorily. Also this type of structure has never been observed by the writer in an iron alloy containing an appreciable amount of carbon.

Written Discussion: By C. Nusbaum, Case School of Applied Science, Cleveland.

On the first reading of Mr. Goss' paper, one receives the impression that it consists of several not very closely related parts. However, if thoughtfully read, it becomes quite evident that the basic or fundamental idea which runs consistently through the entire paper is that of the "block structure" of grains. It is this particular concept and phase of the paper which the speaker wishes to discuss. The discussion, for the sake of ease and clearness of presentation, may be treated under three headings: (a) the presentation of the historical background of the idea of "block structure," and a clear-cut statement of what the authors mean by the term; (b) the relationship between "block structure" and "radial asterism"; and also (c) the importance of the idea of "block structure" as related to the physical properties of crystalline materials.

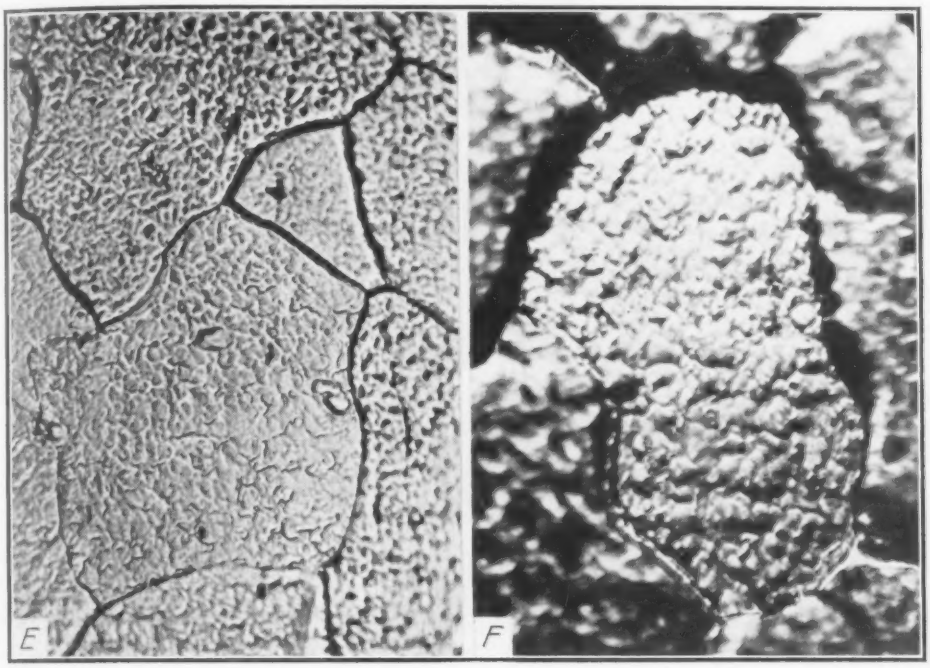
(a) The historical background of the terms, "mosaic," "fibre" and "block structure" of the grains of crystalline materials. Shortly after the now classical experiment of Bragg on the reflection of X-rays from rock salt crystals,



Figs. A and B—Silicon Steel (Strip Stock 0.025 Inch Thick) 3 Per Cent Silicon, Annealed 1600 Degrees Fahr. (871 Degrees Cent.) × 1000. Deep Etch in 2 Per Cent Nital. (A) and (B) are same Area but Different Focus.

Figs. C and D—78 Permalloy (Strip Stock 0.014 Inch Thick) 78 Per Cent Nickel, 22 Per Cent Iron, Annealed 2012 Degrees Fahr. (1100 Degrees Cent.) Air Cooled. × 2000. Deep Etch in Merica's Reagent. Shows Fine Grain Structure in Austenitic Material.

Darwin observed that the reflections were considerably more intense than should be expected from theoretical considerations. The explanation offered was that any crystal of this material was not perfect, but rather built up of small units, or blocks; that these individual units, or blocks, corresponded more closely to our idea of a perfect atomic lattice than the crystal aggregate of which they were a part; also, that the atomic planes of one block may lack parallelism from that of its neighbors by a few, or even several, minutes. To such a structure of single crystal aggregate, he applied the term, "mosaic."



Figs. E and F—Armco Iron (Sheet Stock 0.125 Inch Thick). × 2000. (E) Deep Etch (F) Very Deep Etch 2 Per Cent Nital. Only One Grain of (F) Could be Brought in Focus at one Time.

Among so-called single crystals, rock salt is perhaps considered the best example of a "mosaic" structure, while calcite perhaps corresponds best to our concept of a perfect crystal; that is, if the single crystal is built up of blocks, the corresponding atomic planes of the adjoining individual blocks may lack parallelism by a few, or several, seconds instead of minutes, as in rock salt. The term "mosaic" has been universally accepted by physicists.

Several years ago, Zwicky's theoretical work on the attractive forces within crystals led him to the conclusion that all so-called single crystals, or grains, are inherently built up of "blocks" or have a "block structure." Further, that some crystals are "structure sensitive," and others "structure insensitive." It was the only way in which he could reasonably account for the large discrepancy between the theoretical value of the tensile strength of a crystal and its experimental value. Within the last few years, in the field of X-rays as well as metallography, there has been increasing evidence in support of the "block structure" idea. Some of these contributions are mentioned in Mr. Goss' paper. Since a discussion of the various contributions is in preparation for

publication at the present time, such further evidence, except one—radial asterism—will not be discussed on this occasion.

It is, I believe, generally conceded that when a metal is plastically deformed, motion takes place within each grain along definite crystal planes. These planes are called "slip planes," or "planes of easiest slip," and are characteristic of the crystal family to which the metal belongs. That such is the case, has been amply shown by the extensive experimental work on single crystals of metals. As the plastic flow continues, the number of "slip planes" increases, so that slip takes place not only along one plane but along many approximately parallel planes. This gives a crystal, or grain, a striated or lamellar appearance. Initially, these planes may be spaced many thousand of atom planes apart, but as the flow continues the planes increase in number and the distance between such planes may be equivalent only to a few hundred, possibly more or less, of atomic planes. It is apparent that some of the many possible parallel atomic planes are inherently more susceptible to slip than others, evidently due to differences in the atomic forces.

If now we consider each one of these layers, or lamellae if you choose, as built up of "blocks," the atomic planes of each "block" making a small angle with the corresponding atomic planes of the neighboring blocks, you have perhaps as clear a picture as can be presented of what we mean by a "block structure." Every grain is inherently built up of "blocks" each of which may be a perfect crystal, or it in turn may be built up of still other but smaller blocks.

If the angular deviation between these inherent blocks is small in magnitude, the resulting structure is called "mosaic." The deviation can only be detected by precision X-ray methods. If the angular deviation is large, and especially if there is a rotation of the "blocks" around an axis normal to the plane of slip, this rotation increasing between succeeding layers, the resulting structure is called a fibrous structure, or texture. Such can be detected and by the Laue method of X-ray analysis. The term "block structure" is general and includes both "mosaic" and "fibrous."

(b) The relationship of "block structure" to "radial asterism." Very often in Laue X-ray photographs the diffraction spots are not well defined, but are definitely elongated, broadened, or both. In many cases the spots seem to have a definite structure, that is, instead of single elongated spots, gradually decreasing in intensity, it is made up of a series of spots of nearly the same intensity and slightly superimposed. This phenomenon is generally known as "radial asterism," since the radial elongation of the spots is generally more pronounced. Unfortunately "radial asterism," from the very beginning, has been considered as due to the curvature of atomic planes, and interpreted by many as indicating a form of "lattice distortion." It is an explanation which is given in most texts on the subject, and found extensively in the literature. This interpretation seems to the writer as unfortunate and erroneous. In a series of stress experiments (both tension and bending) on single and multicrystalline materials, the writer has never observed "radial asterism" present until the sample under stress has definitely passed the elastic limit of the material. This means that "radial asterism" is not observed until plastic flow has begun, or, in other words, until slip has taken place between adjoining

lamellae (preferably termed "blocks") along slip planes. Plastic flow is a necessary condition for the presence of "radial asterism," and, as will be indicated later, "lattice distortion" plays no part.

When grains are stressed within the elastic limit, "lattice distortion" undoubtedly takes place, but the distortion produced is of such small magnitude that the Laue method will not detect it. More sensitive methods are necessary. Such elastic strains indicate their presence by causing diffraction lines to become diffuse. "Lattice distortion" may also be caused by a constituent being in solid solution.

Explanation of "radial asterism." In a master's thesis presented by Goss to Case School of Applied Science in May of 1928 on "An X-Ray Study of Plastic Deformation of Iron Wires," occurs this significant statement: "Each radial streak must be due to a large number of (1 1 0) planes, all of which lie in the general direction of the streak but which differ slightly in orientation from crystal fragment to crystal fragment." In a later publication (TRANSACTIONS, A.S.M., October 1934, page 623) by Mr. Goss and the speaker, entitled "Grain Distortion in Metals," they state that in addition to the displacement already mentioned, rotation of the respective lamellae (grain fragments or "blocks") around an axis perpendicular to the slip planes undoubtedly takes place. The reflection of an X-ray beam from such lamellae ("blocks") will produce a diffraction maximum (1) radially elongated when the plane of the beam coincides with the plane of lamellae, and (2) transversely elongated when the plane of the beam is at right angles to the plane of the lamellae ("blocks"). That such a rotation takes place is quite necessary for a satisfactory explanation of the mechanism of preferred orientation. Also, "a distorted grain produced during heat treatment may be defined as one which, during its state of growth, has not grown freely, but has been disturbed in such a manner as to cause motion or displacement (both translation and rotation) of its individual parts ('blocks') along slip planes. This motion along slip planes of a grain, accompanied by rotation of the lattice fragments ('blocks'), or by the rotation of the lamellae ('blocks') themselves relative to each other, is indicated by the presence of 'radial asterism' in its Laue photographs." Still further, "To distort a growing grain during heat treatment it is then only necessary that the internal forces due to grain growth exceed the elastic limit of the material for the temperature at which grain growth is taking place." Zwicky has shown from theoretical considerations that when the rate of grain growth is sufficient, plastic deformation should occur. "Radial asterism" in the Laue X-ray photographs of heat treated metals can thus only be interpreted as indicating that the grains have not grown freely, and that it is built up of unit blocks which lack parallelism with each other. It gives us an indication of the manner and conditions of grain growth. If the individual spots are sharply defined, then the individual units, or blocks, approach parallelism. The grain approximates a perfect crystal.

When a metal is mechanically treated so as to be increasingly stressed beyond its elastic limit, an increasing number of slip planes is set up along which the "blocks" of the respective grains glide both in translation and rotation. With continued treatment, these blocks break up into repeatedly smaller units. This process is indicated by "radial asterism." The conclusion

that "radial asterism," whether present in mechanically deformed or heat treated metals, indicates that the individual grains are built up of smaller units which during the process have been translated and rotated relative to each other, is exceedingly difficult to avoid, and indeed seems inevitable.

(c) The importance of the idea of "block structure" as related to the physical properties of crystalline materials. It seems desirable to make a comment in regard to work hardening. The present day tendencies in regard to work hardening are unusually clearly outlined in his paper on "Single Crystals of Metals and Plastic Deformation." There is no doubt but that in single crystals the slip takes place along definite planes and in definite directions in these planes, or that there is no rotation of the lamellae or adjoining layers relative to each other. However, in multicrystalline materials where the orientation of the grains is a random one, the forces acting on any grain are very complex in nature and do not in any manner correspond to the controlled external forces always acting in single crystal experiments. It is well recognized that the early stages of plastic flow in multicrystalline materials is one of fragmentations, the greatest amount taking place in the region from 0 to 10 per cent reduction in area. The smallest of these blocks are of the order of magnitude of 2×10^{-5} cms. while many others are considerably larger. It is difficult to see how under such circumstances there would be no rotation of the blocks relative to each other, especially since in the later stages of fragmentation there are already evidences of preferred orientation. It is during this stage of fragmentation that "radial asterism" is most marked.

As already stated, the dimensions of the block units vary greatly in value on fragmentation. From theoretical considerations of potential, it can be readily shown that their lattice parameters should be larger than for the large units. There is considerable experimental evidence in favor of such change in parameter, and also symmetry. The presence of these smaller ones associated with larger ones along slip planes may serve as a keying agent, and thus explain increase in strength in cold working. It is a suggestion worthy of consideration.

It is the opinion of the speaker that X-rays will prove to be just as useful in the study of the earlier stages of plastic flow as they have been in the study of preferred orientation. However, it will be necessary in some cases to very definitely improve the technique of the methods used.

Written Discussion: By A. Goetz, California Institute of Technology, Pasadena, California.

The facts and interpretations concerning the problem of plastic deformation and allied phenomena in metal crystals which are treated in this paper are most interesting also from the standpoint of the physicist. They prove again that crystalline plasticity cannot be described without the assumption of the existence of an approximately periodic subdivision of the crystal. It is well known, on the other hand, that in spite of international symposia and discussions, physicists and crystallographers have not arrived at an agreement as to the existence, the nature and the cause of a macroscopic subdivision of the crystal. The nature of the difficulty lies obviously in finding a cause for a configuration in the crystal which renders it thermodynamically less stable than an uninterrupted structure. In view of the fact that the experimental evidence which can be interpreted in favor of the existence of a "block" structure

in a crystal is unfortunately always of a circumstantial nature (as we never know whether or not the crystal under observation is really nondeformed), one has so far to accept the "blocks" as a descriptive working hypothesis.

It can be assumed as an established fact that under conditions of perfect absence of deformation the "block" structure inside of a crystal is absent and that it appears either by externally applied plastic deformation or by certain deforming tendencies during crystallization (presence of impurities, local supercooling, presence of segregating components). As one cannot assume that the "block" is created out of the crystal continuum by such deformation of a perfect crystal, one must assume its latent existence in the case of a perfect crystal. This implies the existence of the "block" boundary in a crystal before the genesis of the "block". The existence of this boundary which does not necessitate a crystallographic discontinuity in the sense of a mosaic structure of the undeformed crystal, seems to be therefore the essential and necessary assumption.

The author has assumed as cause for the boundary the process of crystallization itself in which a formation of molecular groups takes place caused by the arrangement of molecules, while still in the liquid state, into the configuration of the later crystal lattice. This process will lead necessarily to the formation of molecular aggregates within the otherwise statistically arranged liquid. It is plausible to assume that the size of these groups crowds around a most probable value and it is necessary for thermodynamic reasons to assume the existence of a surface between the arrayed group and its nonarrayed surroundings. This surface is then a place of a small discontinuity which as such is "frozen" into the later rigid crystal.

Arguing along such lines one arrives at a definite group size (i.e. a definite block size after deformation) and also at the fact that in an undeformed crystal the discontinuities have not to appear by necessity. These assumptions are supported by a large variety of experiments in which a change of the physical properties of a metal crystal appears when its size is smaller than a certain critical size (i.e. the group size). It is interesting to note that almost all such experiments result in critical sizes of the same magnitude, i.e. $0.1 - 1\mu^3$, that is, to molecular aggregates of 10^9 to 10^{11} molecules.

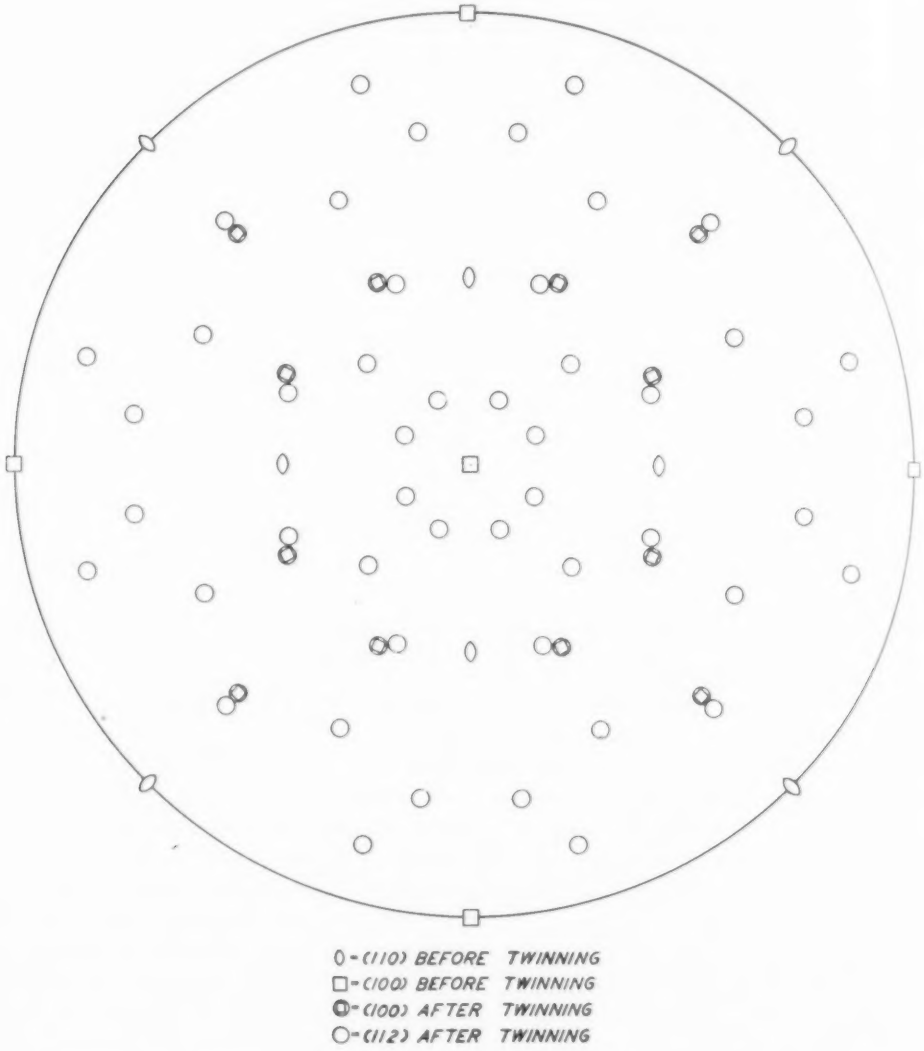
Generally speaking, it seems to be that most of the mechanical properties of metals and probably also a number of their electromagnetic qualities used for practical purposes depend to a much larger extent upon the nature of the group boundaries in the crystal than on properties of the crystal lattice itself, and also that the mechanical properties of metals are controlled to a large extent by changing the nature of the group boundaries by cold and hot work as well as by the addition of foreign atoms in small concentrations.

Generally speaking, a better understanding of the conditions at the internal surface of the crystal seems even more important than the behavior of the grain boundaries. If thus the closer study of the asterism on X-ray diagrams proves to yield a method for the quantitative investigation of such problems as Mr. Goss' work suggests, a most valuable step forward has been made.

Written Discussion: By C. S. Barrett, Carnegie Institute of Technology, Pittsburgh.

This is in reply to the question of the position of the spots that would be predicted on Mr. Goss' twinning theory of recrystallization. If the rolling

structure is one described in reference 20 as [1 1 0] parallel to the rolling direction and a {1 0 0} plane in the rolling plane, with a scatter of about 10 degrees from this ideal orientation, then twinning on {1 1 2} planes would produce a structure giving spots (with the beam normal to the rolled sheet)



Stereographic Projection of a Cubic Crystal and the Twins That Can Form in It.

at approximately the following angles from the rolling direction: 30 and 60 degrees on the (1 1 0) ring; 0 and 90 degrees on the (2 0 0) ring; and 0, 36, 54 and 90 degrees on the (2 1 1) ring. These angles were determined from the stereographic projection of a cubic crystal and the twins that could form in it (see the accompanying plot). On the other hand, if the other orientation described in the reference is present (with [1 1 0] parallel to the rolling plane and a {2 1 1} plane parallel to the rolling plane) the position of the spots can be calculated by rotating the above projection 35 degrees about the [1 1 0]

direction as an axis so as to bring a {2 1 1} plane parallel to the rolling plane. It is then seen that there would be reflections at approximately the following angles from the rolling direction: 32, 57, 59 degrees on the (1 1 0) ring, 12, 76 and 90 degrees on the (2 0 0) ring and 9, 29, 43, 74 and 90 degrees on the (2 1 1) ring. In these calculations it was assumed that any pole lying within about 10 degrees from the corresponding reflection circle would reflect, for a scatter of about 10 degrees is postulated in the references cited in this discussion.

Neither of these calculated patterns agree with the patterns Mr. Goss has observed. In checking over the calculations by which Mr. Goss concluded that twinning would account for the recrystallization texture, I find them in error because they are based on the assumption that {2 1 1} planes in a cubic crystal can be perpendicular to each other, and also the assumption that a plane lying parallel to the incident beam of X-rays can reflect these rays, both of which are incorrect.

Written Discussion: By J. Kunz, Professor of Physics, University of Illinois, Urbana, Illinois.

This paper deals with metals and alloys. X-rays have been used for the investigation. Some of the important results are as follows:

1. The metallic crystals are subdivided into smaller units, called blocks. Asterism and diffuse diffraction spots do not indicate internal strain, but blocks which are somewhat irregularly arranged by rotation and translation.

2. Hardening is explained by the block structure. Plastic flow consists in the displacement of blocks. Some slip planes within the block, which have been displaced by cold working, for instance, are stronger than the ultimate strength of the metal. Moreover through plastic deformation the blocks break up into smaller particles ranging from colloidal size up to the block itself. The new units increase the resistance to plastic flow, i.e., hardening. By tempering, the colloidal particles within the block grow somewhat larger and more uniform, an effect which decreases the hardness again.

3. The block structure affects the physical properties of the metallic grains and crystals. For instance, in silicon steel the blocks are large and perfectly aligned in each grain, therefore silicon steel has good magnetic properties, while the same steel without silicon has smaller blocks of poorer alignment and therefore poorer magnetic properties. I was surprised to find on page 981 the statement that X-ray diffraction proves that the lattice dimensions as well as the cubical symmetry are changed slightly as the particles within the displaced blocks become smaller and smaller.

Written Discussion: By G. Ansel, Carnegie Institute of Technology, Pittsburgh.

The suggestion that slip lines in iron have a saw-tooth structure resulting from slip on {1 1 0} only is not new and has been considered at length in the discussion to the paper by Barrett, Ansel and Mehl, "Slip, Twinning and Cleavage in Iron and Silicon Ferrite," presented at this convention. We have found no microscopic evidence to support this theory and we have shown that it is contrary to accepted principles and requires some very improbable assumptions.

Author's Reply to Written Discussions

The written discussions by Professor Nusbaum, Professor Goetz and Mr. Cramer are in themselves excellent contributions to the concept of block structure. The excellent deep etch figures presented by Mr. Cramer are convincing support for block structure. There are many, however, who do not feel that one is justified in calling deep etch figures "blocks," since different reagents give different deep etch effects, and therefore the exact size of the blocks cannot be determined. The deep etch figures certainly prove the existence of the "block" structure, and for the moment that is all we are interested in. On the other hand, the selective deep etching of the same crystal by various reagents points to the "fine structure" properties of the "blocks" rather than against their existence. Bitter patterns should also prove useful in investigations on "block" structure. Bitter patterns (6) also show evidence of block structure, but the patterns are quite complicated.

As pointed out in the paper, X-ray methods of crystal analysis can be successfully employed in investigations on "block" structure, since asterism and diffuse diffraction spots are related to variations in the disposition of the blocks, rather than internal strain.

Professor Goetz conceives "blocks" as "groups" and personally I think that "groups" is a better way to describe this, inasmuch as it tells us why blocks exist. A paragraph from Professor Goetz's paper, page 70 (7a), explains the origin; to quote. "Before entering the solid state in crystalline form a metal has to go through a paracrystalline interphase in which molecular groups are formed, inside which the geometric configuration is already close to that of the solid lattice. The size of these groups is a physical constant of the crystal and lies between 10^{-14} to 10^{-15} (cm.^{3}). The boundaries of these groups may give rise to the discontinuities in the solid crystal."

X-ray "reflection" diagrams of metals only a few degrees above their melting point should show the existence of "groups."

Professor Goetz also points out the "block" structure is inherent even in so-called perfect crystals, and so he shows the same viewpoint as the writer.

It would be interesting to obtain Laue diagrams of crystals, having asterism and free of it, and then obtain Bitter and deep etch diagrams. These would tell an interesting story.

In 1934 Professor Nusbaum and the author presented experimental evidence which proved that asterism is not a phenomenon confined only to cold-worked metals, but is also found on X-ray diagrams of heat treated metals. This can only mean that the blocks were displaced on the glide surfaces during heat treatment.

Block structure plays an important part in the magnetic properties of ferromagnetic materials. In magnetic materials having excellent properties, one finds the blocks to be larger, and the alignment quite perfect. This I believe would support Professor Goetz's and Professor Nusbaum's viewpoint that internal surface is more important than the behavior of the grain boundaries.

I am pleased to note that Dr. Howard Smith is of the opinion that in view of what has been given in this paper, it will be necessary to reinvestigate some previous work. His suggestions are worth while. In fact they must

be considered if definite progress is to be made, and previous concepts satisfactorily cleared up.

The reason for the intense preferred fiber pattern in the T direction is explained entirely from a distribution basis. Apparently very little cold work is required to orient the grains in this direction, however in the N and P directions, the spread remains large, and so the orientation of the grains does not show up until considerable (plastic deformation) has been given to the metal. The distribution of the grain fragments (blocks) has a profound effect on the physical properties in these directions. In one instance a small hot-rolled strip was cross rolled (alternately) so that the P and T diagrams were practically alike, and certain physical properties were found to be the same. Therefore the "block" distribution has a marked effect on the physical properties in various directions in the surface of the sheet.

Dr. Barrett and Mr. Ansel are of the opinion that the explanation I have given regarding shear failure at 60 degrees in tensile test specimens in strip form is invalid, since I did not consider the possibility of {2 1 1} and {1 2 3} as being shear planes. In my explanation I only considered the {1 1 0} planes as shear planes.

The reason for the explanation I did give is simply this: the maximum shear theory requires the fracture to occur at 45 degrees but the failures always occur at an angle of about 60 degrees. It is certainly remarkable to find how close to the 60 degrees the fractures are.

The shear failure at 60 degrees can be satisfactorily explained purely from the lattice theory, and the orientation of the lattice fragments when a steel strip is gradually elongated in the tensile machine.

Let us consider what happens in an annealed specimen in which the grains are initially oriented at random. As the strip is gradually elongated in the tensile machine the grains are fragmented into displaced block systems as already explained. X-ray examination of the fracture reveals the fact that the grain fragments are oriented with the [1 1 0] direction coinciding with the axis of tension. This requires two other (1 1 0) planes to make an angle of 60 degrees with the axis of pull. Since the angle of fracture also is approximately 60 degrees, then the (1 1 0) planes must be the shear planes. The X-ray diagram also shows that the diffraction maximum are arcs (about 10 degrees). This adequately accounts for the slight deviation from 60 degrees and irregular fracture of the test piece.

At the present time no one has ever shown that {1 2 3} and {2 1 1} planes are slip planes in commercial steels. Even in the case of pure iron, when these planes have been reported as glide planes, the literature indicates that the whole situation is quite controversial. At any rate the point I wished to make was that the 60 degrees shear failure could be nicely accounted for from crystal structure, since the maximum shear theory failed to explain it.

Dr. Barrett also criticized the interpretation I gave several years ago in which I made an attempt to explain the recrystallization structure as due to twinning on the {2 1 1} planes. In answering Dr. Barrett I asked him to show what the distribution of the diffraction maxima must be on the (2 1 1) ring in order to account or justify a twinning structure.

This he has done, and if I understand him correctly, twinning would be accounted for if the diffraction maxima on the (2 1 1) ring are found at 0, 36, 54 and 90 degrees.

By direct measurement, Fig. 8 (page 411 "effect of annealing on the grain structure of extremely hard rolled steel ribbon"—Sept. 1929 TRANSACTIONS, American Society for Steel Treating) we find that the maxima appear at 0, 35, 55 and 90 degrees as required by Dr. Barrett's own calculations.

If the cubic is twinned on the (2 1 1) planes one will find that (2 1 1) planes can make an angle of 90 degrees with each other, in fact the distribution on the (2 1 1) ring (0 and 90 degrees) verifies this (See Fig. 8, page 411).

Naturally these twins are not oriented ideally, but are distributed statistically, and the limits can be easily determined by setting a specimen at various angles to the X-ray beam.

I also wish to express my gratitude to Professor Kunz for the concise resume of the facts which he felt were of importance. It includes some of the points which I wanted to express; that "asterism" and diffuse diffraction spots are not indicative of internal strain (lattice distortion) but show that the blocks are somewhat irregularly arranged by rotation and translation; the blocks have a "fine structure" which controls the physical properties, for example the hardness.

In conclusion I wish to thank everyone for their discussions which have added much by bringing to light a better conception of the subjects discussed.

The remaining nine papers which were presented as part of the Symposium on the Plastic Working of Metals will be published in the March 1937 issue of TRANSACTIONS.—Editor.

TECHNICAL PROGRAM AND REPORTS OF OFFICERS AMERICAN SOCIETY FOR METALS—18th ANNUAL CONVENTION, CLEVELAND, OCTOBER 19 to 23, 1936

FOR purposes of record and for the benefit of members who were not in attendance at the Eighteenth Annual Convention of the Society, held in Cleveland, October 19 to 23, 1936, the Technical Papers Program and Reports of Officers presented at the annual meeting are herewith published in full.

MONDAY, OCTOBER 19

Morning Session—Statler Hotel Ball Room—10:00 A. M.

Joint Chairmen—H. M. Boylston and R. Sergeson

Research Session

The Retarding Effect of Certain Metallic Elements on Graphitization, H. A. Schwartz, H. H. Johnson and C. H. Junge, National Malleable and Steel Castings Co., Cleveland.

Cadmium-Silver-Copper Alloys for Engine Bearings, C. F. Smart, Pontiac Motor Co., Pontiac, Mich.

High Speed Motion Picture Showing Behavior of Quenching Mediums During Quenching, I. N. Zavarine, Massachusetts Institute of Technology, Cambridge, Mass.

The Diffusion of Hydrogen through Nickel and Iron, W. R. Ham, Pennsylvania State College, State College, Pa.

Afternoon Session—Public Auditorium—Club Room B—2:00 P. M.

Joint Chairmen—E. J. P. Fisher and H. J. Stein

Heat Treating Session

Notes on Continuous Gas Carburizing, R. J. Cowan, Surface Combustion Corp., Toledo, Ohio.

Differential Hardening by Induction, M. A. Tran, Park Drop Forge Co., Cleveland, and W. E. Benninghoff, Ohio Crank Shaft Co., Cleveland.

The Continuous Heat Treating of Cold-Rolled Strip, N. P. Goss, Cold Metal Process Co., Youngstown, Ohio, and T. B. Bechtel, Electric Furnace Co., Salem, Ohio.

Afternoon Simultaneous Session—Public Auditorium—Club Room A 2:00 P. M.

Joint Chairmen—H. E. Handy and N. I. Stotz

Research Session

Effect of Titanium on Some Cast Ferrous and Nonferrous Metals, J. A. Duma, Norfolk Navy Yard, Portsmouth, Va.

Effect of Titanium on the Hardness and Microstructure of Heat Treated 18 Per Cent Chromium Steel Ingots, R. E. Bannon, Titanium Alloy Manufacturing Co., Niagara Falls, N. Y.

The Production of Flakes in Steel by Heating in Hydrogen, R. E. Cramer, University of Illinois, Urbana, Ill.

Public Auditorium—Club Room B—4:30 and 8:00 P. M.

Educational Lectures

Physical Testing of Metals and Interpretation of Test Results, Prof. H. D. Churchill, Case School of Applied Science, Cleveland.

Radiography in Principle and in Practice, Dr. K. R. Van Horn, Aluminum Company of America, Cleveland.

TUESDAY, OCTOBER 20

Morning Session—Statler Hotel Ball Room—10:00 A. M.

Joint Chairmen—G. B. Waterhouse and A. E. White

Grain Size Session

Some Effects of Small Additions of Vanadium to Eutectoid Steel, J. G. Zimmerman, R. H. Aborn and E. C. Bain, U. S. Steel Corporation, Kearny, N. J.

Influence of Aluminum on the Normality of Steel, G. R. Brophy and E. R. Parker, General Electric Co., Schenectady, N. Y.

A Study of the Effect of the Aluminum Addition on the Structure of a Quenched Carbon Steel, H. W. McQuaid, Republic Steel Corp., Massillon, Ohio.

Afternoon Session—Public Auditorium—Club Room B—2:00 P. M.

Joint Chairmen—A. L. Boegehold and L. W. Kempf

Physical Properties Session

Effect of Overload on the Fatigue Properties of Several Steels at Various Low Temperatures, H. B. Wishart and S. W. Lyon, University of Illinois, Urbana, Ill.

Austenitic Stainless Alloys: Their Properties and Characteristics, V. N. Krivobok, R. A. Lincoln and R. Patterson, Jr., Allegheny Steel Co., Brackenridge, Pa.

Slip, Twinning and Cleavage in Iron and Silicon Ferrite, C. S. Barrett, G. Ansel and R. F. Mehl, Carnegie Institute of Technology, Pittsburgh.

Dimension Changes of Tool Steels During Quenching and Tempering, Ainar Ameen, Uddeholms A.B., Sweden.

Public Auditorium—Club Room B—4:30 and 8:00 P. M.

Educational Lectures

Physical Testing of Metals and Interpretation of Test Results, Prof. H. D. Churchill, Case School of Applied Science, Cleveland.

Principles and Apparatus for X-Ray Diffraction Analysis, Dr. K. R. Van Horn, Aluminum Company of America, Cleveland.

WEDNESDAY, OCTOBER 21

Morning Session—Statler Hotel Ball Room—9:30 A. M.

ANNUAL MEETING OF THE A.S.M.

1936 Edward DeMille Campbell Memorial Lecture presented by J. P. Gill, Vanadium-Alloys Steel Co., Latrobe, Pa.

Chairman—A. T. Clarage

Afternoon Session—Public Auditorium—Club Room B—2:00 P. M.

Joint Chairmen—A. A. Bates and C. H. Herty, Jr.

Melting Session

The Magnetic Properties of a Series of Basic Open-Hearth Slag Samples, B. A. Rogers and K. O. Stamm, United States Bureau of Mines, Pittsburgh.

Basic Open-Hearth Slag Control, Earnshaw Cook, American Brake Shoe & Foundry Co., Mahwah, N. J.

A New Tool for the Control of Quality Steel Making, G. T. Motok, Republic Steel Corp., Cleveland.

Equilibrium in the Reaction of Hydrogen with Iron Sulphide in Liquid Iron and the Thermodynamics of Desulphurization, John Chipman, American Rolling Mill Co., Middletown, Ohio, and Ta Li, China.

Public Auditorium—Club Room B—4:30 and 8:00 P. M.

Educational Lectures

Physical Testing of Metals and Interpretation of Test Results, Prof. H. D. Churchill, Case School of Applied Science, Cleveland.

Industrial Applications of X-Ray Diffraction Methods, Dr. K. R. Van Horn, Aluminum Company of America, Cleveland.

THURSDAY, OCTOBER 22

Morning Session—Statler Hotel Ball Room—9:30 A. M.

Joint Chairmen—E. S. Davenport and O. E. Harder

Symposium on the Plastic Working of Metals

Laws and Fundamentals of Plastic Deformation, A. V. deForest, Massachusetts Institute of Technology, Cambridge, Mass.

Metallic Single Crystals and Plastic Deformation, S. L. Hoyt, A. O. Smith Corporation, Milwaukee, Wis.

Creep Characteristics of Metals, C. L. Clark and A. E. White, University of Michigan, Ann Arbor, Michigan.

Interpretation and Use of Creep Results, J. J. Kanter, Crane Co., Chicago.

Damping Capacity: Its Variation and Relation to Other Physical Properties, G. R. Brophy and E. R. Parker, General Electric Co., Schenectady, N. Y.

Morning Simultaneous Session—Statler Hotel—Salle Moderne Room
10:00 A. M.

Joint Chairmen—R. F. Mehl and K. R. Van Horn

X-Ray Session

An X-Ray Study of Preferred Orientations in Pure Cold-Rolled Iron-Nickel Alloys, D. McLachlan, Jr., and W. P. Davey, Pennsylvania State College, State College, Pa.

The Application of X-Ray Diffraction to the Study of Fatigue in Metals, C. S. Barrett, Carnegie Institute of Technology, Pittsburgh.

X-Ray Diffraction Studies of Distortion in Metals, G. L. Clark, University of Illinois, Urbana, Ill., and M. M. Beckwith, Guide Lamp Co., Anderson, Ind.

Afternoon Session—Public Auditorium—Club Room B—2:00 P. M.

Joint Chairmen—H. J. French and V. N. Krivobok

Symposium on the Plastic Working of Metals

Elastic Properties and Their Relationship to Strength and to Strain Hardening, M. F. Sayre, Union College, Schenectady, N. Y.

The Effect of the Shape of the Test Piece Upon the Energy Needed to Deform Materials in the Single-Blow Drop Test, O. W. Ellis, Ontario Research Foundation, Toronto.

Hot Working, Cold Working and Recrystallization Structure, N. P. Goss, Cold Metal Process Co., Youngstown, Ohio.

Factors Relating to the Production of Drop and Hammer Forgings, Adam M. Steever, Columbia Tool Steel Co., Chicago Hts., Ill.

Afternoon Simultaneous Session—Public Auditorium—Club Room A
2:00 P. M.

Joint Chairman—J. P. Gill and I. C. Matthews

Research Session

Further Study of a High Carbon-High Chromium Tool Steel, W. H. Wills, Ludlum Steel Co., Dunkirk, N. Y.

The Importance of Boundary Attack in the Etching of Steel Specimens, B. L. McCarthy, Wickwire Spencer Steel Co., Buffalo, N. Y.

Recovery of Cold-Worked Nickel at Elevated Temperatures, Erich Fetz, Wilbur B. Driver Co., Newark, N. J.

Public Auditorium—Club Room B—4:30 P. M.

Educational Lecture

Physical Testing of Metals and Interpretation of Test Results, Prof. H. D. Churchill, Case School of Applied Science, Cleveland.

7:00 P. M.—Ball Room, Hotel Statler

ANNUAL BANQUET OF THE A.S.M.

FRIDAY, OCTOBER 23

Morning Session—Statler Hotel Ball Room—10:00 A. M.

Joint Chairmen—E. F. Ross and H. J. Stagg

Symposium on the Plastic Working of Metals

Hot Press and Upset Forging, J. H. Friedman, National Machinery Company, Tiffin, Ohio.

Extrusion of Metals, D. K. Crampton, Chase Brass & Copper Company, Waterbury, Conn.

Cold Heading—Bolts, Rivets, Nails, R. H. Smith, Lamson and Sessions Company, Kent, Ohio.

Cold Forming Processes—Drawing Rods and Bars, J. E. Beck, Jones & Laughlin Steel Corp., Pittsburgh.

Morning Simultaneous Session—Statler Hotel Lattice Room—10:00 A. M.

Joint Chairmen—J. L. Burns and M. L. Frey

Testing Session

Conversion of Tensile Test Data from One Form of Test Piece to Another, E. J. Janitzky, Carnegie-Illinois Steel Co., Chicago.

The Behavior of Some Low Alloy Steels in the Single-Blow Drop Test, O. W. Ellis, Ontario Research Foundation, Toronto.

The Fracture of Carbon Steels at Elevated Temperatures, A. E. White, C. L. Clark, University of Michigan, Ann Arbor, Mich., and R. L. Wilson, Timken Steel & Tube Co., Canton, Ohio.

Afternoon Session—Statler Hotel Ball Room—2:00 P. M.

Joint Chairmen—C. L. Clark and D. F. McFarland

Symposium on the Plastic Working of Metals

Cold Drawing Seamless Steel Tubes, Horace Knerr, Consulting Metallurgical Engineer, Philadelphia.

The Cold Rolling of Mild Steel Sheets and Strip, Anson Hayes and R. S. Burns, American Rolling Mill Co., Middletown, Ohio.

Some Factors Affecting the Plastic Deformation of Sheet and Strip Steel and Their Relation to the Deep Drawing Properties, Joseph Winlock and R. W. E. Leiter, Edward G. Budd Mfg. Co., Philadelphia.

Cold Working of Hollow Cylinders by Auto-Frettage, N. E. Woldman, Eclipse Aviation Corp., E. Orange, N. J.

Afternoon Simultaneous Session—Statler Hotel Lattice Room—2:00 P. M.

Joint Chairmen—G. T. Williams and W. P. Wood

Physical Properties Session

Investigation of Fatigue Strength of Axles, Press Fits, Surface Rolling, and Effect of Size, T. V. Buckwalter and O. J. Horger, Timken Roller Bearing Co., Canton, Ohio.

Endurance of Gear Steels at Two Hundred Fifty Degrees Fahr., A. L. Boegehold, General Motors Corp., Detroit.

Physical Properties of Axle Shafts, H. B. Knowlton, International Harvester Co., Chicago.

Statler Hotel Ball Room—4:30 P. M.

Educational Lecture

Physical Testing of Metals and Interpretation of Test Results, Prof. H. D. Churchill, Case School of Applied Science, Cleveland.

ANNUAL ADDRESS OF THE PRESIDENT

Eighteenth Annual Convention, Cleveland, October 19, 1936

ROBERT S. ARCHER, *President*

OUR Society is in excellent condition. That our activities are worth-while is shown by our record membership, by the willingness of so many of our members to devote valuable time to the work in which we are mutually engaged, and by the support we are receiving from the metal industries. Our ability to continue these activities is indicated by a healthy financial condition.

During the past year, the Board of Trustees has held three meet-

ings in Cleveland on November 1, 1935, July 10, 1936, and October 20, 1936. Committees and individuals have been appointed, and the Trustees desire to express their appreciation to those who have helped to carry on the work of our Society. We are particularly indebted to Professor H. D. Churchill for his series of lectures on "The Physical Testing of Metals", and to Dr. K. R. Van Horn for his lectures on "X-Ray Analysis", both of these series being presented during the present Convention.

The President's Medal was presented to Past-President W. H. Phillips at the annual banquet during the Convention in 1935.

The Henry Marion Howe Medal, awarded for the paper of highest merit published in TRANSACTIONS during the year, was awarded for the year 1935 to Dr. T. D. Yensen of the Westinghouse Electric and Manufacturing Company, and Mr. N. A. Ziegler of the West Penn Electric Company, for their paper entitled "Magnetic Properties of Iron as Affected by Carbon, Oxygen and Grain Size", published in TRANSACTIONS in June 1935. No medal will be awarded in 1936, because of a change in plan to permit a longer period of time for consideration and selection. The award for the paper of highest merit published during 1936 is to be made in 1937.

The Campbell Memorial Lecture, established in 1925 by the Board of Trustees in honor of the memory of Edward DeMille Campbell of the University of Michigan, was presented in 1935 at the Convention in Chicago by Harry W. McQuaid, metallurgist, Republic Steel Corporation, Massillon, Ohio. His lecture was entitled "The Importance of Aluminum Additions in Modern Commercial Steels," and was published in full in the December 1935 issue of TRANSACTIONS.

Albert Sauveur Achievement Award

The Albert Sauveur Achievement Award was established by the Board of Trustees in 1934. The award rests entirely in the hands of the past presidents of the Society, who constitute the Albert Sauveur Award Committee.

Dr. Sauveur, in whose honor the award was established, was made the first recipient at the annual banquet in New York in 1934, and Dr. Zay Jeffries received the award in 1935.

This year I am pleased to report the committee has selected W. D. Chapin. The presentation will take place at the annual banquet Thursday evening, October 22, 1936.

President's Bell

At the close of the administration of R. M. Bird in 1926, he established an award to be presented annually by the president to be known as the President's Bell, which would be awarded in recognition of the outstanding activities of the chapter as well as its contribution to the success of the Society.

During the past years it has been awarded to the following chapters:

1926—Lehigh Valley	1931—Hartford
1927—Golden Gate	1932—Montreal
1928—Philadelphia	1933—Cleveland
1929—Los Angeles	1934—Cleveland
1930—New Jersey	1935—Chicago

This year, as frequently in the past, it has been a problem to select one from among many deserving Chapters. The selection has been made however, and the bell is hereby awarded to the Peoria Chapter.

Those who have received their copies of the 1936 edition of the Handbook have recognized that it is a work of increased value and scope, presented in a form of greater convenience and beauty. We wish to express our appreciation to the Handbook Committee, to the many contributors and reviewers, and to J. E. Donnellan, permanent Secretary of the Committee.

A communication has been received from our Rochester Chapter transmitting a resolution passed by that Chapter in which is expressed a desire for an increased proportion of papers of a technological or practical nature. This desire has been recognized for many years, and a constant effort has been made to secure more papers of this type, and at the same time of high quality. At the meeting of the Board on October 20, 1936, steps were taken to secure additional material of this kind by a new plan, details of which will soon be communicated to all local chapters.

It has not been practicable for me to visit all of our 40-odd chapters during the past year, but I have had the pleasure of visiting many of them. I have found increased membership, increased educational activities, and in general a high degree of enthusiasm.

One of the distinguishing characteristics of our Society has been our willingness—yes, eagerness—to hear the experiences and thoughts of young men, of relatively inexperienced men, or of older men who do not easily express themselves. No member need hesitate to dis-

cuss freely the subject of any of our meetings, local or national. We all respect the interest and sincerity of those who enter earnestly into these discussions, whether or not we agree. We want new facts, whether or not such facts appear to be in accord with previously accepted beliefs. By seeking new facts from all of our membership, by stimulating free discussion and argument rather than mere questions, and by drawing upon the energy and enthusiasm of our younger members, let us seek to perpetuate the vitality which has been ours in the past, and which is ours today.

REPORT OF THE TREASURER

October 21, 1936

W. P. WOODSIDE, *Treasurer*

The cash position of the Society as shown below is approximately the same as a year ago, even though we have just purchased over $30,000.00 of additional securities, our receivables are higher by $4000.00 as the volume of advertising and sale of books and services have increased materially. Inventory has gone up about $2500.00 as additional books have been published. The account with a restricted bank of approximately $3700.00 remains the same although the prospects for receiving this sum in full are considerably brighter.

The market value of securities held on August 31st of last year and still in the portfolio on August 31, 1936 has not increased materially but the sale of some holdings and purchase of additional securities has improved the position substantially.

The total amount invested is.....$166,367.69
The total present market value is..... 152,850.94

To offset this difference we have a reserve for market depreciation of $25,000.00. Improved market valuation has enabled us to reduce this reserve during the past year by $5,000.00 from the previous figure of $30,000.00 and it appears that present values warrant another substantial reduction of this reserve.

Sales and calls of securities

$5000.00 Texas Corporation 5's
5000.00 Swift & Company 5's
5000.00 Youngstown Sheet & Tube Co. 5's

Repurchased

$5000.00 Texas Corporation 3½'s
6000.00 Youngstown Sheet & Tube Convertible 3½'s

New Purchases

$5000.00 American Gas & Electric Co.
5000.00 Anaconda Copper Mining Co.
5000.00 Public Service Co. of Northern Illinois
5000.00 Pacific Gas & Electric Co.
5000.00 Virginia Railway Co.
200 shares Massachusetts Investors Trust
200 shares Incorporated Investors

Our portfolio now shows investment in

Government bonds	$ 54,000.00
Utilities	50,000.00
Railroads	25,000.00
Industrials	21,000.00
Investment Trusts	10,000.00

With possibility of inflation the Investment Trust shares were purchased as a mild hedge.

Our audited statement of income and expense for the fiscal year ended August 31, 1936 is as follows in total:

Income	$231,486.46	
Expense	201,288.41	
Net Gain		$ 30,198.05
Net Gain from Society Activities	$ 19,238.34	
1935 Convention	10,530.31	
Appreciation on bonds	429.40	

Advertising income from Metal Progress again increased by approximately 40 per cent over the previous period and we were able to show an excess income over expense on this activity of $7500.00.

The sale of books published by the society increased substantially during the past year and helped in the showing of a substantial increase in surplus.

The membership total and of course, dues receipts, continued to increase during the last fiscal year and in addition to a new high total of members, added approximately a $10,000.00 increase in dues receipts to our previous gross income. The healthy condition of the membership is, of course, the foundation of the society and determines the success of all other activities which rest on this foundation.

CONDENSED AUDITED BALANCE SHEET
AMERICAN SOCIETY *for* METALS

As of August 31, 1936

ASSETS		
CASH		
On Hand	$ 400.00	
On Deposit	37,361.25	$ 37,761.25

SECURITIES		
U. S. Government Bonds	$ 54,250.65	
Other bonds	112,117.04	
Accrued interest	1,647.24	
	$168,014.93	
Less—Reserve	25,000.00	143,014.93
Accounts Receivable	$ 12,523.13	
Less—Reserve	527.16	11,995.97
Inventory	$ 25,746.65	
Less—Reserve	3,000.00	22,746.65
OTHER ASSETS		
Travel Advances, Employees' Accounts, etc.	$ 3,787.89	
Deposits in Restricted Banks	3,666.43	7,454.32
Office Furniture and Fixtures less Depreciation		6,250.00
Prepaid Expense (Convention, Metal Progress, Insurance)		18,600.56
		$247,823.68
LIABILITIES and SURPLUS		
Accounts Payable		$ 3,451.04
Reserves		45,000.00
Advance 1936 convention receipts		42,418.62
Surplus		156,954.02
		$247,823.68

The above from Balance Sheet prepared by Ernst & Ernst, Certified Public Accountants.

INCOME AND EXPENSE
AMERICAN SOCIETY *for* METALS

For the year ended August 31, 1936

INCOME			
"METAL PROGRESS"			
Advertising		$79,729.75	
Subscriptions		2,384.91	
Reprints		1,943.23	
Sales		165.80	$84,223.69
MEMBERSHIPS			
Membership dues	$82,541.49		
Sustaining exhibitors' dues	4,975.00	$87,516.49	
Less: Apportionment of dues to chapters		33,078.15	54,438.34
1935 Convention—Chicago			63,765.82
Books published and purchased for resale			17,447.43
Interest earned			5,067.62
"Transactions"—bound copies, subscriptions, etc.			4,290.39
Discount earned			1,283.77
Profit on Sale or Redemption of bonds			433.40
Chapter technical service			169.40
Miscellaneous			43.00
General Reprints			289.60
"The Review"—advertising and subscriptions			34.00
TOTAL INCOME			$231,486.46
EXPENSE			
AS SHOWN BY EXHIBIT			
"Metal Progress"		$76,723.08	
1935 Convention—Chicago		53,235.51	
Books published and purchased for resale		18,838.44	
"Transactions"—bound copies, subscriptions, etc.		16,274.20	
General expense		8,686.71	
Chapters		2,674.94	
"The Review"		2,335.70	
Secretary's Office		13,713.79	
Accounting department		5,004.53	
Directors' expense		1,577.92	

National committees	1,034.25		
President's expense	421.96	$200,521.03	
Lectures, medals, etc.		443.13	
General index		188.93	
Books for library		112.68	
Reprints purchased for resale		22.64	
TOTAL EXPENSE			$201,288.41
NET INCOME			$ 30,198.05

The above from Income and Expense sheet prepared by Ernst & Ernst, Certified Public Accountants.

In addition to the figures shown in the audited reports we have received reports from the chapters showing a total of approximately $56,000.00 in assets held by them, which is an increase over last year's total of $4,000.00 and indicates an equally healthy financial position for the local chapters.

ANNUAL REPORT OF THE SECRETARY

October 21, 1936

W. H. EISENMAN, *Secretary*

The American Society for Metals, on October 1, 1936, had a total membership of 7691. Of this number 6808 or 88.5 per cent were the member classification, 492 or 6.4 per cent were the sustaining classification, 368 or 4.8 per cent were the junior classification, with 23 honorary members.

On October 1 last year the Society had a total membership of 6880. This shows an increase for the year of 811 or 11.8 per cent.

MEMBERSHIP 1920-1936

Year	Total Members	Increase	Per Cent Increase
1920	1724	...	....
1921	2168	444	25.7
1922	2313	145	6.7
1923	2746	433	18.7
1924	3026	280	10.2
1925	3363	337	11.1
1926	4000	637	18.9
1927	4653	653	16.3
1928	4844	191	4.1
1929	5615	771	16.0
1930	6138	523	9.3
1931	5761	—377	—6.15
1932	4468	—1293	—22.4
1933	4470	2	0.045
1934	5443	973	21.8
1935	6880	1437	26.4
1936	7691	811	11.8

Publication Committee

The Publication Committee for the year 1936 was made up of the following: W. P. Sykes, Chairman; Ray T. Bayless, Secretary; H. E. Ardahl, A. A. Bates, C. L. Clark, E. S. Davenport, C. N. Dawe, M. A. Grossmann, W. E Harvey, G. V. Luerssen, D. F. McFarland, O. W. McMullan, R. F. Mehl, N. I. Stotz, and Joseph Winlock.

Throughout the year and up to the present time the Committee has reviewed 54 papers, of which 48 have been approved, 5 rejected, and 1 withdrawn by the author.

The meeting for the year 1936 was held in Cleveland at the National Offices, June 19th.

The following was made a part of the record of this Committee meeting:

A discussion of the papers for the Symposium on the Plastic Working of Metals, and it was agreed that the Symposium should be divided into four half-day sessions, dividing the papers as equitably as possible between the four sessions, so that the maximum amount of time would be provided for presentation and discussion.

There was a general discussion on the subject of papers for the 1937 Convention. After discussing the matter of Symposia and the various subjects that could be discussed as Symposia, some dozen or more subjects were considered. As a result of this discussion, it was resolved that it was the desire of the Committee to hold a Symposium in 1937 at the time of the annual meeting and it was moved and seconded that the subject of the Symposium would be "Carburizing and De-Carburizing."

Transactions

There have been published since the last annual meeting of the Society four quarterly issues of TRANSACTIONS, namely, December, March, June and September. The number of pages totals 1083. 40 papers were printed, together with the discussion.

Metal Progress

METAL PROGRESS has completed the most successful year of its history, both as an instrument for disseminating technical information to members, and as a business venture. In common with other trade and technical magazines, losses were experienced during the

depression years, but these were held to nominal amounts by careful management. Each issue of 1936 has cost less to produce than the income from advertising and subscriptions. The October issue, now in your hands, breaks all records for size, revenue, and diversity of interest. The consolidated balance sheet for the six years of operation shows a comfortable and encouraging surplus.

In accordance with a resolution by the Trustees, President Archer appointed early in his administration an Editorial Board, which has held several meetings and has aided the Editor in securing a well-balanced, interesting and valuable reading section in each issue. Notable improvements and innovations have been:

First, A series of cover engravings and designs that has set a new high in artistry;

Second, Portraits and brief biographies of important personages in metallurgy, both technologists and industrialists;

Third, Pictorial stories, wherein a metallurgical operation such as forging is portrayed by a series of interesting photographs rather than by words.

The Board of Trustees acknowledge the debt it owes to the many unselfish members of the Society who have contributed such excellent articles to METAL PROGRESS.

The Review

During the past year THE REVIEW has been published as a bimonthly. The November 1935 and January and March 1936 issues each contained eight pages. In May it was decided that a slight enlargement of the paper would permit a more complete reporting of chapter activities as well as a more interesting and readable presentation. The May issue was therefore expanded to twelve pages, a size that will probably be used during the coming season.

The September issue was devoted almost entirely to the 1936 National Metal Congress. It contained twelve pages and had a circulation of 17,000 among the members of all societies and divisions cooperating in the Congress.

From November 1935 to date (six issues) 2192 column-inches were devoted to editorial matter and 548 to advertising. The amount of THE REVIEW space devoted to reports of chapter meetings was 1012 column-inches.

Preprints

Thirty papers were presented before the 17th annual convention held in Chicago last year (1935). For this year's convention 50 have been scheduled for presentation, 38 of which have been preprinted and distributed to those members requesting them. The total number of pages of preprints for 1935 was 688, whereas the total number of preprint pages for 1936 is 933. The number of requests for preprints this year is the largest in our history, well over 1000 having been received.

Books

During the past year the Society has published the following books: "Principles of Heat Treatment" by M. A. Grossmann, which consists of the series of lectures presented at the Chicago convention; "Engineering Alloys," by Norman E. Woldman and Albert J. Dornblatt—this book gives the production, manufacturer, and physical properties of over 8000 alloys; and "Steel Physical Properties Atlas" by Charles Newman Dawe.

There is now in preparation for publication the lectures on Physical Testing of Metals by Prof. H. D. Churchill, which are being presented at this convention.

Metals Handbook Committee

Metals Handbook Committee has completed its work on the new 1936 edition of the A.S.M. Metals Handbook, which is now being distributed to the members of the Society. The personnel of Metals Handbook Committee, whose responsibility it was to organize and compile the 1936 edition, was as follows:

W. C. Chancellor
R. L. Dowdell
H. J. Fischbeck (Term Expired 1935)
J. P. Gill
J. H. Hall (A.S.T.M. Representative)
W. D. Halsey (Resigned 1935, A.W.S. Rep.)
S. L. Hoyt
C. H. Herty, Jr., Chairman
H. B. Knowlton (Term Expired 1935)
F. B. Lounsberry (Term Expired 1935)
G. C. Riegel
E. C. Smith (Term Expired 1935)
A. M. Steever
Jerome Strauss (Chairman, Nonferrous Data Sheet Committee, Inst. Metals Div.)
J. E. Donnellan, Secretary

During the last year, Metals Handbook Committee had one meeting. The meeting was held in the National office of the Society on January 30, 1936. At this meeting the Committee completed its plans for the final preparation of the 1936 edition of the Handbook.

In order that all articles and practices would be revised for the new Handbook, the Committee reappointed all former subcommittees that had served on various occasions. All authors were also requested to review their articles and make whatever revisions were required. As a result 620 members of the Society have served on 41 committees and 360 authors have participated in the preparation of the present edition. In addition to this, every article prepared for Metals Handbook Committee was reviewed by at least two to four members. Therefore, approximately 2350 members of the Society have participated in the preparation of Metals Handbook.

After the completion of a Handbook, it is the policy of Metals Handbook Committee to discharge from active duty all subcommittees. Therefore, all Handbook committees are inactive for the time being. When Metals Handbook Committee has its first meeting for the year of 1937 plans will again be worked out and the subcommittees reappointed with assignments for new work.

The Handbook has been produced in the larger 6 x 9 inch size instead of the former Handbook size. This new size is much more convenient as a reference book and has enabled the Society to issue the Handbook as a single unit. If the former size had been retained, it would have been necessary to issue the book in two volumes, or print it on paper of a much lighter weight. The present edition contains a total of 1540 pages, covering 335 subjects, 97 of which were new for the 1936 edition.

Because of the large volume of data and information published in the Handbook, the index has been made more extensive. The members will find the index more complete and satisfactory than in any previous edition.

Metals Handbook contains approximately 2500 references to the metallurgical literature. For this present edition the Society has had all these references checked by the Engineering Library in New York to make sure that they were accurate and complete so that they would be of maximum service to the members and readers of the Handbook.

In the preparation of the present edition, the same Ionic type face was selected, as used in the last edition. This particular type face was developed by one of the large type manufacturers for ease and comfort of reading, but at the same time to conserve space which makes it quite appropriate for Handbook purposes.

The same plan of arranging and organizing the various articles

and sections of the Handbook was followed as in the former edition. This method, you will recall, was proposed a few years ago by R. S. Archer. As it has proved so satisfactory, the committee has continued this scheme for the present edition. This plan classifies the subjects under seven major heads, but with the various sub-divisions the present edition contains 52 different divisions.

The International Acetylene Association co-operated with the Society in the preparation of the section in the Handbook on Welding and Cutting of Metals. The Association appointed a Welding Committee composed of both gas and electric welding experts. The personnel of the Welding Committee was as follows:

C. W. Obert, *Chairman*	E. W. P. Smith
E. V. David	S. C. Osborn
L. B. Meeker	A. G. Wikoff

As Welding is such an important subject, the readers of Metals Handbook will find this section prepared by the International Acetylene Association a valuable addition.

The following divisions and number of articles in each division of the Ferrous Section of the Handbook are as follows:

General Data	10	Testing	14
Testing—General	19	Melting	10
Heat Treatment—General	2	Mechanical Working	7
Welding and Cutting of Metals	2	Heat Treatment	30
Processes, Methods and Equipment	9	Machining and Grinding	3
Definitions, Trade Names, and Weights	4	Surface Treatment	18
Constitution of Ferrous Alloys	17	Applications	4
Properties of Iron and Iron Alloys	31	TOTAL	180

As in other editions of Metals Handbook, the Nonferrous Section was prepared by the Nonferrous Data Sheet Committee of the Institute of Metals Division of the American Institute of Mining & Metallurgical Engineers, through a co-operative arrangement with the American Society for Metals. The Nonferrous Data Sheet Committee has completed a very successful and important program, and has greatly developed the Nonferrous Section of the Handbook. We wish to express our sincere appreciation to the Nonferrous Data Sheet Committee and to the members of the American Institute for Mining and Metallurgical Engineers, who have so willingly given their whole-hearted co-operation in the preparation of this important section of Metals Handbook. The personnel of the Nonferrous Data Sheet Committee was as follows:

Jerome Strauss, *Chairman*
L. W. Kempf
N. B. Pilling
Arthur Phillips (Term Expired Feb., 1935)
W. L. Fink
W. E. Remmers
W. G. Schneider
E. E. Schumacher
R. S. Archer (Term Expired Feb., 1935)
E. M. Wise

The 1936 edition of Metals Handbook is much wider in scope than any of the previous editions and contains much additional information on the technology, properties, and applications of ferrous and nonferrous metals. Metals Handbook will undoubtedly take its place as one of the important metallurgical reference books. This edition of the Handbook demonstrates the unselfish interest and co-operation of the entire membership of the Society, since the members have given freely of their time and technical information in the compilation and preparation of this work.

The Exposition

1928—Philadelphia	$62,460.00
1929—Cleveland	62,420.00
1930—Chicago	45,650.00
1931—Boston	39,360.00
1932—Buffalo	22,890.00
1933—Detroit	29,440.00
1934—New York	42,100.00
1935—Chicago	62,170.00
1936—Cleveland	70,000.00

This year's exposition is the largest metal show in A.S.M. history—or for that part in the history of America. 217 exhibitors have placed in the Public Auditorium for your study a most attractive, satisfying and complete display. These shows are profitable to them and to you. You owe it to yourself and to the manufacturer to give him your attention and interest. You will see gathered together there, under one roof, all that is new and worth while in the industry. Exhibitors are deserving of the thanks and commendation of the entire membership for their splendid co-operation and helpful interest.

Educational Activities

The splendid educational work carried on by the chapters is a service to which the A.S.M. may point with pride. Practically every chapter carried on some type of educational work during the past year. These courses provided a review and new information for the members and permitted the chapter to demonstrate that it was fulfilling its functions of serving the metal industry and the metal man.

There were three courses available from the national office; a series of eight lectures prepared by Dr. A. Allan Bates of Case School of Applied Science on the "Fundamentals of Ferrous Metallurgy," while the second was the series of lectures presented by Dr. Marcus A. Grossmann at the Chicago convention last year on "The Principles of Heat Treatment," and the third was the course prepared by Mr. James P. Gill covering the subject of "Tool Steels."

These lectures in printed form and with lantern slides were used by 21 chapters during the past year.

The chapter executive committees, as well as those upon whom the responsibility for the presentation of these educational lectures fell, are entitled to receive the most sincere thanks and appreciation of the Society.

The annual meeting of the A.S.M. is replete with scientific papers, but to advance the practical side of the national convention, there was inaugurated three years ago a series of lectures upon a practical subject by an expert in that line. It was recognized that the chapter meetings afford a splendid opportunity for the presentation of practical talks, as is indicated by the fact that not 5% of the talks thus presented are ever published. Most of the chapter speakers report they had talked extemporaneously and upon fundamentals, and that the material was not suitable for any of the publications of the Society.

It is further recognized that the annual meeting of the Society is the only opportunity afforded to those who are engaged in researches and investigations for the presentation of their findings and a public recognition of the work the authors have accomplished.

The board of trustees, however, recognized that many attending the convention were interested in the fundamental facts of the metal science as well as the many practical items of interest displayed in the exposition.

So it was decided that practical lecture courses should be held during the week of the convention. This policy has been in practice for two years with remarkable success so that this third year a series of lectures on "The Physical Testing of Metals and Interpretation of Test Results" is being presented by Profes or H. D. Churchill, associate Professor of metallurgy, Case School of Applied Science.

In addition to the above, Dr. Kent R. Van Horn, research metallurgist of the Aluminum Co. of America, is presenting three lectures on "X-Ray Analysis."

The success of the A.S.M., a technical society, is similar to the success of an industrial organization. An industrial organization depends for its success upon its product and the men behind that product.

The product must be good, the men alive, alert, progressive, unselfish and hard working. With these attributes, an industrial organization is sure and assured of success.

And so with the A.S.M. Through the 18 years of its existence, its product has met the approval of an ever increasing number of men —until today over 8000 members (300 of whom joined in the first 15 days of October) work together for the advancement of its aims and objects.

As an industrial organization depends on its personnel, so does the A.S.M. depend upon its national committees and chapter officers and executive committeemen. And such loyal, alert, progressive, unselfish and hard working men they are! Even though the fundamental idea back of the A.S.M. was sound, it never in a hundred years could have achieved its present respected and admired position without their aid. To one who was present at the birth of the Society, who assisted in teaching the infant to walk, who observed the growing pains of its early childhood, it is a thrill that plays vibrant music on one's heart strings to know and see how others too have learned to love, honor and work for the A.S.M. Their work is not for a day, or a week or a month or a year, but year after year after year.

In recognition of that vast army of workers throughout the membership, the board of trustees has invited the six secretaries of chapters who have accepted the responsibility of that office for ten or more years to be the "guests of honor" and guests of the Society at the 18th annual convention. They are:

H. E. Handy—10 years as secretary of the Boston chapter.

I. C. Matthews—10 years as secretary of the Rochester chapter.

C. G. Peterson—11 years as secretary of the Rhode Island chapter.

H. L. Walker—13 years as secretary of the Pittsburgh chapter.

T. N. Holden—15 years as secretary of New York chapter.

Alexis Caswell—16 years as secretary of the North West chapter; elected in 1920 at the organization meeting.

Members—here in the combined service of these six men you observe 75 years of faithful, untiring effort for the A.S.M. These men are representative of hundreds of other members who have and

are working for the advancement of the Society, with the same splendid, unselfish spirit.

It is no wonder then that the Society grows by leaps and bounds, that it constantly increases its services to the members, that its financial position improves from year to year, that it holds the largest metal exposition in America, that its publications—METAL PROGRESS, TRANSACTIONS and METALS HANDBOOK—bring letters of praise from all over the world. It is no wonder, and why? Just pull aside the curtain and behold a hive of busy members—8000 of them—joining as one in a labor of love—the building of the A.S.M.

ELECTION OF OFFICERS

ROBERT S. ARCHER, *President*

I have the pleasure to report that in accordance with the constitution of the A.S.M. the President received prior to March 1 of this year suggested candidates for the Nominating Committee from all the local chapters of the Society entitled to make nominations. From the list of eligible candidates suggested, I appointed the following Nominating Committee of seven:

H. B. Northrup—Penn State—chairman
B. L. McCarthy—Buffalo
E. F. Ross—Cleveland
M. E. Cummings—Syracuse
E. A. Thomas—Muncie
R. W. Woodward—Hartford
C. R. Whittemore—Montreal

The names of the members of the Nominating Committee were published in the March REVIEW.

The Nominating Committee responded to the call of the President and held a meeting in Syracuse during the third full week in May and reported immediately the selection of the following candidates for the offices indicated.

For president for one year—E. C. Bain
For vice president for one year—G. B. Waterhouse
For secretary for two years—W. H. Eisenman
For trustee for two years—R. L. Wilson
For trustee for two years—O. W. Ellis

The report of the Nominating Committee was published in the May REVIEW.

I have been notified by the Secretary that no additional nominations were received prior to July 15 for any of the vacancies occurring

on the board of trustees. Consequently, the nominations were closed. I shall now call upon the secretary to perform his duty as prescribed by the constitution.

W. H. EISENMAN, *Secretary*: Since no additional nominations were received prior to July 15, 1936, for any of the vacancies on the board of trustees of the American Society for Metals, I herewith, in compliance with the provision of the constitution, cast the unanimous vote of all of the members of the Society for the election of the following candidates:

For president for one year—E. C. Bain
For vice president for one year—G. B. Waterhouse
For secretary for two years—W. H. Eisenman
For trustee for two years—R. L. Wilson
For trustee for two years—O. W. Ellis

R. S. ARCHER: Since all the provisions of the constitution have now been performed, I herewith declare the candidates named to have been duly and unanimously elected for the term and office specified, and the term of office of each officer just elected shall begin on the day following the close of this annual meeting.

The Annual Banquet

The American Society for Metals was host to more than eight hundred members and guests at its annual banquet held Thursday evening, October 22, in the Ballroom of the Statler Hotel, Cleveland. Many executives of important steel producing and steel consuming companies were in attendance. Seated at the speakers' table were the following: C. E. Denney, president, Erie Railroad; George Brooks, vice president, Chesapeake and Ohio Railroad; B. F. Fairless, president, Carnegie-Illinois Steel Corp.; R. J. Wysor, executive vice president, Republic Steel Corp.; Walter Mathesius, manager of operations, Carnegie-Illinois Steel Corp. and trustee A.S.M.; E. C. Bain, assistant to vice president, U. S. Steel Corp. and president-elect of A.S.M.; G. B. Waterhouse, head department of metallurgy, Massachusetts Institute of Technology and vice president-elect of A.S.M.; W. P. Woodside, vice president, Climax Molybdenum Corp. and a founder member and treasurer of the A.S.M.; Honorable Harold H. Burton, mayor of City of Cleveland; Zay Jeffries, General Electric Co., a past president of the Society; W. R. Chapin, E. C. Atkins Co., Sauveur Medallist; B. F. Shepherd,

manager Rock Drill Dept., Ingersoll-Rand Co., and past president of A.S.M.; Dr. Albert Sauveur, Harvard University; H. V. Kaltenborn, principal speaker of the evening; J. P. Gill, chief metallurgist, Vanadium-Alloys Steel Co., who delivered the Edward DeMille Campbell Memorial Lecture for 1936; Alexis Caswell, secretary of North West Chapter since 1920; T. N. Holden, secretary of New York Chapter since 1921; H. L. Walker, secretary Pittsburgh Chapter since 1923; Howard E. Handy, secretary of Boston Chapter since 1925; Irving C. Matthews, secretary of Rochester Chapter since 1926; R. L. Kenyon, supervising research engineer, American Rolling Mill Co., and trustee of A.S.M.; S. C. Spalding, metallurgist, American Brass Company, and trustee of A.S.M.; O. W. Ellis, director, Ontario Research Foundation and trustee-elect of the A.S.M.; Theodore Barker, president, Accurate Steel Treating Co., and a founder member of the A.S.M.; Kent R. Van Horn, research metallurgist, Aluminum Company of America, and author of series of lectures on X-Ray Analysis at this Convention; H. D. Churchill, associate professor of metallurgy, Case School of Applied Science, and author of the series of lectures on Physical Testing of Metals and Interpretation of Test Results delivered at this Convention; H. P. Croft, metallurgist, Chase Brass Company, and chairman of Cleveland Chapter; H. E. Brown, metallurgist, W. S. Tyler Company, and secretary of Cleveland Chapter; Robert S. Archer, Republic Steel Corp., and president of A.S.M.; W. H. Eisenman, secretary, American Society for Metals.

President R. S. Archer expressed the appreciation of the Trustees and the Membership of the American Society for Metals to the Institute of Metals, and the Iron and Steel Division of the American Institute of Mining and Metallurgical Engineers, The American Welding Society and the Wire Association and the American Society of Mechanical Engineers for their hearty co-operation in making this Eighteenth National Metal Congress such an outstanding success.

President Archer then introduced each of those seated at the speakers table to the assembled guests.

Presentation of Past President's Medal

In continuing President Archer said:

"It now becomes my privilege and duty to present the Past President's Medal to Benjamin Franklin Shepherd. For many years

Mr. Shepherd has worked tirelessly in the best interests of this Society. He has been a member of the Society since its inception and has served his own Lehigh Valley Chapter loyally and faithfully. He has served as a member of the Board of Trustees of the National Society for a period of two years and then as Vice-President and President and Past President for one year each. The Past President's Medal is now awarded to him in sincere appreciation of his splendid services to the organization."

Albert Sauveur Achievement Award

"It is my pleasure to introduce to you Dr. Zay Jeffries, the recipient of the Albert Sauveur Achievement Award for 1935, who will present the candidate for this honor for 1936." The committee of award consists of all Past Presidents of the Society.

In presenting Mr. W. R. Chapin for the Sauveur Achievement Award Dr. Zay Jeffries said:

"Mr. President, Ladies and Gentlemen:

"I have been delegated by the president and the committee, consisting of all living past presidents of the American Society for Metals, to present the candidate for the 1936 Sauveur Achievement Award. Our selection is Mr. W. R. Chapin, of Indianapolis.

"Following is the citation: 'For his fundamental work in connection with two-stage quenching of steel, which has explained many mysteries of heat treatment results, has stimulated and guided many important researches, and has led to the clarification of theory and improvement in practice of steel-treating.'

"Mr. Chapin's contribution was first presented in Cleveland before the local Chapter during the Winter 1921-1922. Several persons in this room had the pleasure of listening to his modest but able report. The paper was published in Vol. 2 of the TRANSACTIONS of the American Society for Steel Treating, page 507. It therefore complies with one of the requirements for the Award, namely, that the contribution be given the acid test of time.

"It would not be possible, even though time were available, to list the researches and the philosophical discussions which have been importantly affected by Chapin's work. Suffice it to say that they are many and varied.

"He confined his report to pieces of steel of such size or composition, or combination of these factors, as would harden all the way through when quenched in oil. By interrupting the quench

from 1475 degrees Fahr. he found a temperature range around 400-500 degrees Fahr. in which the critical change, though normally occurring around 1300 degrees Fahr. on slow cooling, had not taken place at all. The steel was soft to the Brinell ball test, was non-magnetic and plastic. It was, however, hard by the file test. He found that the steels were quite stable if kept in this temperature range and that the real hardening took place on further cooling toward room temperature. He noted the normal volume contraction resulting from cooling to this so-called 'hardening range,' but a marked expansion and return of magnetism accompanying the hardening, all taking place below about 400 degrees Fahr.

"Here are a few quotations from Chapin's paper:

'. . . . . there is less distortion and less ultimate change of volume if these steels are allowed to cool slowly from the hardening range to the cold than when allowed to cool in the quenching bath as in common practice.'

'Since this expansion begins and is most rapid at the thinnest section and in the region farthest from the center of the steel, it naturally follows that any quenched steel such as punches, dies, cutters, etc., should be withdrawn from the quenching medium in the hardening range and placed at once in a medium sufficiently hot to permit the martensite to form slowly and completely and uniformly but not hot enough to start the decomposition of the martensite into troostite or sorbite.'

'It may be stated as a truth that a steel properly treated in the hardening range, providing of course that the steel is of the correct analysis and has been made right, will be less liable to rupture and will show a minimum of deformation.'

'Soft spots in oil-hardening steels which harden solid can be detected in the hardening range with absolute certainty by the use of a magnet. The writer has demonstrated this beyond a reasonable doubt. It should be strongly emphasized that a steel taken from the quenching bath in the hardening range should be placed at once in a medium of the proper temperature. Steels withdrawn in the hardening range are very liable to rupture if allowed to cool in the air. The reason is perfectly obvious.'

'It seems almost paradoxical that these steels exhibit in the hardening range a low Brinell and at the same time are practically file hard and can be deformed readily. Similar observations have been made on a variety of other steels which harden solid in oil, such steels containing about 1.00 per cent tungsten . . . . . '

'The appearance of the Brinell indentation in the hardening range is peculiar in that it is not clear cut but there is a rounding off entirely around the periphery. This appearance is characteristic of all the steels tested in the hardening range.'

"By way of comment on the last two statements perhaps the difference between the file and indentation tests is nowhere better exemplified. The file removes the metal in contact with it which, in Chapin's experiments, had been cooled by the file itself a sufficient amount to produce hardening. On the other hand, the Brinell ball deforms metal far removed from the contact area. The steel in contact with the ball had no doubt been hardened by cooling, but that at some distance from the contact area remained warm and hence soft.

"In an art like that of steel-treating, which measures its age in thousands of years, it is difficult to allocate the credit for the many contributions. However, in the present instance, after giving generous credit to his predecessors and followers, there is abundance remaining for Chapin.

"Portevin and Garvin, in France, published a paper in 1919 which contains part of the information discovered independently by Chapin. This research is a brilliant one. The paper contains 90 pages. Thirty-three pages and twenty-three curves and figures of the paper are used to describe the new and special apparatus and the calculations leading to the estimation of cooling rates. The apparatus was so special and the procedure so involved that it is doubtful, with the single exception of French and his associates, whether their technique for the determination of cooling rates has ever been excelled. This is another way of saying that the methods and apparatus were special and not available to metallurgists generally. Furthermore, Portevin and Garvin dealt essentially with the uninterrupted quench and determined no physical properties except at room temperature.

"Chapin not only added much that they did not disclose but he confirmed by direct observation certain conclusions which they had arrived at by indirect means. Furthermore, and more important, he used only the tools of the ordinary metallurgical laboratory. When he published his results the whole metallurgical world was put in a position to carry on without added equipment, strange technique or complicated calculations. Perhaps the best known extension of Chapin's work and his methods, in essence, are the beautiful researches of Bain and his associates relating to the interrupted quench.

"Mr. President, it gives me great satisfaction to present as a candidate for The Sauveur Achievement Award for 1936—Mr. W. R. Chapin."

In accepting the Albert Sauveur Achievement Award, W. R. Chapin spoke as follows:

"Dr. Sauveur, Dr. Jeffries, Mr. President, Ladies and Gentlemen:

"I shall remember this moment all the rest of my life. Thoughts rush in. There is a feeling of humbleness mixed with very great joy, humbleness in the presence of this distinguished gathering.

"The real joy of research, it seems to me comes from the realization that when such work is turned over to industry it ultimately results in the making of better products at less cost so that everybody benefits.

"When any group of American citizens sees fit to elevate one of that group to a position of honor and responsibility, that citizen should so conduct himself that in the future there shall be no change of judgment on the part of those who did the elevating.

"Our dear Dr. Sauveur, this great honor is deeply appreciated and there also comes a sobering thought. While it may be very thrilling to see a player break through the line and go fifty yards for a touchdown, the truth is, that such a play could not have happened except for the careful training of the coach, the right signals by the quarterback, and perhaps most of all because of the silent, almost unobserved players in the line.

"It is in like spirit that this Award is accepted, not for myself alone, but for all those who have gone before, and particularly for the men in the mines, in the mills, and in the hardening shops, who are sweating away at their work, for the Indianapolis Chapter, for my business associates and my family.

"I thank you from the bottom of my heart."

Following this presentation Mr. H. V. Kaltenborn, noted radio commentator, lecturer and principal speaker of the evening presented a very entertaining address on the present crisis which is taking place in Spain.

Statement of the ownership, management, circulation, etc., required by the acts of congress of August 24, 1912, and March 3, 1933, of TRANSACTIONS of American Society for Metals, published quarterly at Cleveland, Ohio, for October 1, 1936, State of Ohio, county of Cuyahoga, ss. Before me, a notary public, in and for the State and county aforesaid, personally appeared Ray T. Bayless, who, having been duly sworn according to law, deposes and says that he is the editor of the TRANSACTIONS of the American Society for Metals, and that the following is, to the best of his knowledge and belief, a true statement of the ownership, management, etc., of the aforesaid publication for the date shown in the above caption, required by the Act of August 24, 1912, as amended by the Act of March 3, 1933, embodied in section 537, Postal Laws and Regulations to wit:

1.—That the names and addresses of the publisher, editor, managing editor, and business managers are: Publisher, American Society for Metals, 7016 Euclid Ave., Cleveland, O.; Editor and Managing Editor, Ray T. Bayless, 7016 Euclid Ave., Cleveland, O.; Business Manager, W. H. Eisenman, 7016 Euclid Ave., Cleveland, O.

2.—That the owner is: The American Society for Metals, 7016 Euclid Ave., Cleveland, Ohio, which is an educational institution, the officers being, President, R. S. Archer; Vice President, E. C. Bain; Treasurer, W. P. Woodside; Secretary, W. H. Eisenman; Trustees: B. F. Shepherd, Walter Mathesius, G. B. Waterhouse, R. L. Kenyon, S. C. Spalding. All officers as above, 7016 Euclid Ave., Cleveland, Ohio.

3.—That the known bondholders, mortgagees, and other security holders owning or holding 1 per cent or more of total amount of bonds, mortgages, or other securities are: none.

4.—That the two paragraphs next above, giving the names of the owners, stockholders, and security holders, if any, contain not only the list of stockholders and security holders as they appear upon the books of the company but also, in cases where the stockholder or security holder appears upon the books of the company as trustee or in any other fiduciary relation, the name of the person or corporation for whom such trustee is acting, is given; also that the said two paragraphs contain statements embracing affiant's full knowledge and belief as to the circumstances and conditions under which stockholders and security holders who do not appear upon the books of the company as trustees, hold stock and securities in a capacity other than that of a bona fide owner; and this affiant has no reason to believe that any other person, association, or corporation has any interest direct or indirect in the said stock, bonds, or other securities than as so stated by him. Ray T. Bayless, managing editor, sworn to and subscribed before me this 1st day of October, 1936.

(Seal) Arthur T. Wehrle, notary public.

(My commission expires January 26, 1938.)

Return Your Old Metals Handbook

The new 1936 edition of Metals Handbook, published by the American Society for Metals, is ready for distribution to all members of the Society in good standing. If you have not received your new 1936 Metals Handbook simply return your old 1933 edition to the National Office of the Society and the new book will be mailed immediately.

The value of Metals Handbook has greatly increased, as this edition contains 334 articles on a variety of metallurgical subjects, totaling 1392 pages of data and information. This new reference volume is now produced in a standard 6 x 9 inch size, which makes the Handbook much more convenient and useful. Why be without it?

All you need to do is to return your old copy of the 1933 edition (now obsolete) to the American Society for Metals, 7016 Euclid Avenue, Cleveland, Ohio, and your copy of the new 1936 edition will be sent free and postpaid to you promptly.